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METALLURGY.—*High-strength cast iron: Appraisal and forecast.*[1] JEROME STRAUSS, Vanadium Corporation of America. (Communicated by J. G. Thompson.)

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The art of metallurgy is indeed old, with its beginnings lying in prehistory many millennia behind us. By contrast, metallography, and more particularly metal microscopy, is young. As a science its age is only a few years. Thus many still living, including myself, can well remember pleasant and profitable associations with many of the pioneers.

A substantial proportion of those recently and even at present engaged in assuring that metallography (i.e., physical metallurgy) shall continue to grow as a science began their mature intellectual activities as physicists, and so it was with Dr. George Kimball Burgess to whose memory this evening is dedicated. On me, as on others, he left many lasting impressions—his scientific approach, his convincing presentation, his easy manner, his understanding consideration of a human problem.

Review of the 53 papers of which he was author or coauthor during his 29 years at the National Bureau of Standards is a reminder of his early interest in the measurement of high temperatures, so closely linked with the thermoelectric properties and the radiation characteristics of metals as well as of the products of their oxidation. How natural then for him to become deeply interested in all the fundamental properties of metals and, subsequent to his appointment as chief of the Division of Metallurgy upon its organization, to utilize his knowledge of these fundamentals, of metallography in its broadest sense, and of pyrometry in the study of practical metallurgical problems. Toward these he developed, if it was not actually latent, an extremely keen appreciation and a great capacity for their solution.

Understandably, lecturers are prone to select subjects for discussion that represent advances in fields in which the one memorialized undertook considerable work or that are closely related to this work. Only one of Dr. Burgess's papers, however, touched the subject of cast iron, although that one dealt with research in a phase of use of the product representing most severe service and a problem even now not completely under control. I may therefore embark upon this presentation without feeling that I am departing completely from the realm of his interest.

We are so surrounded by the products of modern technology and as technologists devote so much attention to the details of today's practice that we are inclined to regard steel and wrought iron as invariably the products of refining of cast iron (i.e., carbon-rich iron) and to forget that direct-reduced iron, in spite of all our efforts to produce it economically in these times, was the foundation of what has been termed the Iron Age. What appears to be a significant exception to this sequence of first wrought iron, then cast iron, is the appearance of objects of cast iron in China a hundred or two hundred years prior to the dawn of the Christian Era. Actually this is not an exception—evidence is accumulating to show that

[1] The sixth George Kimball Burgess Memorial Award Lecture to the Washington Chapter, American Society for Metals, on February 21, 1955.

this iron was first reduced from ore in crucibles and then recarburized beyond the range of steels, again in crucibles. In many, if not all, ancient lands early iron objects appear as wrought meteoric iron or direct-reduced iron of malleable quality resulting from the smelting of local ores at low temperature and with limited, if any, fusion. Especially is this true in the Near East. In Europe also the bloomery came long before the "Stückofen," which was the forerunner of the blast furnace. Only when the height of this "Stückofen" was increased above the initial 10 to 15 feet, and the reduced product remained for a longer time in contact with the fuel, did it become possible to obtain a molten high-carbon product consistently, and to produce cast forms deliberately rather than accidentally and occasionally. Thus began European cast iron—some 14 or 15 centuries after its development in China and by a completely different approach.

Why this historical survey? It is a natural question. The recitation is solely a means of recalling that cast iron, known since the second or surely the first century B.C. in China and since the early fourteenth century A.D. in Europe, has attracted the attention of those concerned with metallography and the science of metals to a lesser degree, and only more recently, than its forbear or its offspring (according to your approach being from East or West), namely, steel, with its age as a molten product of only 200 years. Assign such reasons as you will, and very many may claim attention, cast iron since the days of serious metallographic investigation has been the laggard—at least with respect to widespread application of new knowledge. Ultimately the forward movement began, so that one finds in the past 30 years a voluminous literature representing the striving of many foundrymen and researchers seeking not only practical improvements but more knowledge of the fundamentals.

Beginning as an engineering material in Western Europe (unfortunately as always with new materials and processes, most prominently in military engineering) cast iron reached rather later into the field of domestic appliances. First came cannon, then shot for the cannon, and afterward firebacks, cooking utensils, and other articles of daily use. But in the growth of the mechanical industries, the metal played a major role. It was accepted for its advantages, but with its limitations, and applied accordingly. It was easy to produce, whether directly from the blast furnace or by remelting and subsequent casting. It lacked ductility but served well under compression loading and presented an excellent bearing surface. Early measurements of strength were no doubt indirect and comprised primarily performance testing to determine adequacy, but it was known as greatly inferior in strength and ductility to steel and to bronze. Prior to about 1920 (no strength measurements of reasonable reliability may be found earlier than about 1880) it is rare in machinable gray iron to encounter a tensile strength value above 40,000 psi, while in most instances values were below 35,000 psi and very many well below this figure.

Until about that time (i.e., 1920) the old and well-worn saying that cast iron was the product of a process was not at all untrue. It was just not complete in that, in addition to the process, the character and the properties of iron were dependent upon its heredity or more precisely the origin and the character of the raw materials entering the process. It was not a question of the chemical composition of pig iron, iron scrap, steel scrap, etc., but principally of the ores and other raw materials from which the pig iron was produced. Thus the blast furnace giving birth to the iron was all-important, and names had greater significance than the analytical chemist's findings, even though the latter were regularly reported and demanded, and definite limits for silicon, manganese, sulphur, and phosphorus were often insisted upon. Nevertheless, if the results of the process were poor it was more common to look toward or change the origin of the iron; second on the list of blameworthy items was the chemical composition and last the process. The latter was nearly blameless and all else had to fit its demands.

Then came a major change. For 30 or 40 years, sporadic use had been made of large proportions of steel scrap in the cupola

charge. The reason was essentially economic, but occasionally it was lack of pig iron. Records of such charges may be found as far back as the 1880's. A side result of this economic pressure was, however, the production of the higher of the then known mechanical properties. The now obsolete terms of, primarily, semisteel and, secondarily, high-test cast iron came into being. The high strength was traceable to lower carbon content or lower carbon equivalent (C + ⅓ Si), although often in the later years of the period another factor had been introduced, namely ferrosilicon and ferromanganese added during melting. The great change, however, was not entirely due to the addition of steel scrap and the low carbon content; of at least equal importance was the very large increase in melting temperature that it stimulated, and subsequently combining both with inoculation through deoxidation.

The possibilities presented for practical improvements and more significantly for research came from the recognition on the one hand of the effects of starting a cycle of solidification and subsequent cooling from a melt fully liquid, of uniform though perhaps higher gas content but devoid of solid nuclei, and on the other hand (though often interrelated) from altering the character and distribution of the graphite and controlling the matrix microstructure. Research was undertaken in great volume, largely by a few captive foundries, a few forward-looking independents, some associations of manufacturers, and the principal alloy producers. A substantial share of the actual work was performed in universities or under consultation with their staffs. Ultimately the effects of heredity were very greatly minimized and the results were sufficient to command a new respect for an ancient product.

Theoretically—and we omit from consideration other elements of the composition—cast iron is an iron-carbon-silicon alloy produced via the liquid state and containing in excess of 1.7 percent of carbon and sufficient silicon to result in adequate castability. The definition is admittedly loose in view of the partial interdependence of carbon and silicon, but it serves the present purpose. The region here under consideration is principally in the range of 2.4 to 3.7 percent of carbon and 0.5 to 4.0 percent of silicon. There is no novelty in the statement that the subject is exceedingly complex, not only because of the range of essential composition, impurities, alloys, and other additives but also because of the great sensitivity of the microstructure and therefore of the properties of cast iron to section thickness and to all the elements of the manufacturing operation—principally raw materials, melting techniques, maximum temperature in the molten state, pouring temperature, and cooling rate. In the three decades just behind us, the desire for understanding and control of these many factors and their interrelation have supplied the driving force in the search for improvement.

While the subject I have selected as just noted is largely limited to irons with not less than 2.4 percent carbon, there is need in view of a few product types to refer to some irons close to the theoretical minimum of 1.7 percent. Hence a statement of coverage comprises:

1. The combined effects of high proportions of steel scrap as raw material, carbon content just above and below the middle of the range and the concomitant high melting temperature extended to a higher level than thereby required.
2. Inoculation of this high-temperature-melted iron through the creation of nuclei for graphite precipitation.
3. The effects of alloying elements.
4. Changes in the pattern of malleable iron production and use.
5. The introduction of irons containing spherolitic graphite.

The treatment is intended to be partly historical, partly technical, and to include comments on some elements of the picture that remain to be filled in, the data for which may reasonably be anticipated.

LOW CARBON IRONS: APPLICATION OF STEEL SCRAP AND SUPERHEAT

Prior to any systematic studies, the recognition of the benefits to strength resulting from low-carbon content due to large steel scrap additions to the cupola charge, and the moderate temperature increase that followed automatically, had been supplemented by

recognition of the value of a completely pearlitic matrix microstructure.

With respect to low-carbon iron via steel scrap, it should be noted that an alternate means of reaching this objective has been patented by Zenzes (*1*) as early as 1905, his process consisting of blowing pig iron to remove silicon, manganese, and part of the carbon and subsequently mixing with a silicon-rich iron melt. However, the history of steel scrap use shows that McPherran (*2*) went beyond earlier charge limits and reported adding as much as 60 percent steel scrap to the cupola as early as 1913. Later (in 1924) Emmel (*3*) patented a process comprising the addition of 50 percent steel scrap along with ferrosilicon and ferromanganese, all in the cupola. However, I can record the interesting experience of watching McKinney at the U. S. Naval Gun Factory carrying on the same procedure as Emmel 4 or 5 years earlier and extending the scrap addition to well over 90 percent. Before the warnings and criticisms regarding quality and control by those eminent authorities Ledebur and Wuest, McKinney had noted the low quality and the great difficulty of predetermining composition, it being beyond our knowledge at that time to predict whether a tap would carry 2.25 or 2.75 percent carbon. But with improvements in both equipment and analytical control, the principle of Zenzes is now or has recently been in operation, and similarly cupola charges high in steel scrap are not uncommon.

With respect to the production of irons with a 100 percent pearlitic matrix microstructure there had also been early accomplishment. Diefenthaler and Sipp (*4*) working for the Lanz Company had developed prior to 1916 their production practice of preheating the molds to temperatures determined by casting cross-section and iron composition; the latter was specified for each cross-section as the *total* of carbon and silicon but with *individual maxima* specified for each of the two elements. The excellent results of the Lanz method ultimately led to simpler procedures involving high-conductivity molds and high cupola temperatures but an important point is that, while high strength was obtained, it was not specifically sought, the objective being to produce the pearlite matrix for its toughness, wear-resistance and dimensional stability.

It is both appropriate and necessary to continue the discussion at this point with some details of the pearlitic matrix and subsequently return to the use of high melting temperatures. In supplementing rule of thumb and progressing from qualitative to quantitative knowledge, the relationship carbon-silicon-cooling rate-and-section size (these last two are not entirely separable) were given early consideration. As a point of beginning I can do no better than quote the concise statement of Schneidewind and McElwee (*5*):

> The iron-carbon-silicon eutectic may solidify from the liquid state in one of two forms under commercial rates of cooling: either as austenite and carbide or as austenite and graphite. At room temperatures, these will be pearlite and cementite and pearlite and graphite respectively. In the absence of alloying elements the mode of silidification into one or the other of the above mentioned forms or in a mixture of the two is determined by (a) the silicon content, and (b) the rate of cooling from the molten state.

The quantitative relationships began with the publications of Honda and Murakami (*6*) in 1923 and Maurer (*7*) in 1924. Their structure diagrams, representing the results of the simplest approach, were based upon a single section size and cooling rate. To establish the type of graph into which each foundry might incorporate its own special conditions of raw materials, process, and range of products, the section size-cooling rate variable was superimposed by the researches of Uhlitzsch and Weichelt (*8*) in 1933, giving rise to the well-known "Modified Maurer Diagram" of Fig. 1. The permissible region for 100 percent pearlite in thicknesses of ¼ inch to 1¼ inches, and that for carbon contents 2.4 to 3.7 percent are shown slightly offset from each other to permit easy identification.

In plain irons, namely those to which alloying elements have not been added, a full pearlitic structure corresponds to a lower limit of tensile strength of 30,000 to 35,000 psi and an upper limit of 45,000 to 50,000 psi. Below this range free ferrite is in

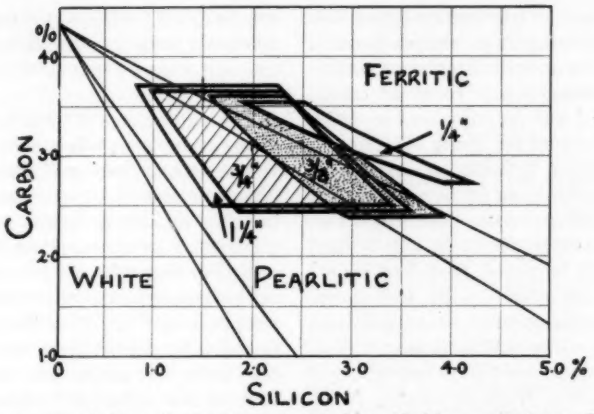

FIG. 1.—Graphic representation of structure of cast from by Maurer (7) for medium section (light lines) with results of Uhlitzsch and Weichelt (8) superimposed for medium to light sections (heavy lines.

evidence, and above it mottling is certain to be encountered. In extending the work of the early investigators on the relationship carbon-silicon-section size, having in mind especially the outlines of the area of 100 percent pearlite, Sipp (*9*) and also Angus, Dunn, and Marles (*10*) set forth the area for sand castings on a carbon equivalent—section thickness plot. The latter authors employed data from many hundred tests, made on irons produced in a large number of British foundries; since British practice with respect to melting temperature tends to be appreciably lower than that in the United States and Germany, it may be assumed in the absence of specific information that these data do not include many, if any at all, resulting from the use of high superheating temperatures. Fig. 2 shows the curves of Angus and coworkers along with those resulting from computations by Schneidewind and McElwee (*5*) in an effort to establigh a formula for computing strength from composition.

Although the term "high-strength cast iron" has not been defined in numerical terms, it is obvious that only the upper portion of the tensile strength range quoted for pearlitic irons has any reason to be placed within its indefinite scope. And while Fig. 2 does reach into the range of low-carbon irons it does not set forth the results that have been achieved by high melting and pouring temperatures, by inoculation, by alloying.

It is in order now to return to the question of steel scrap in the cupola and the effects of increasing melting temperatures. In the long history of this practice, with the attendant increased temperatures and increased strengths, one may find many interesting examples. Among them is one quoted by Pfannenschmidt (*11*) concerning comparisons of cylinder blocks made in 1907 using practices then in vogue, with identical articles produced from the same cupola in 1932 with cast iron and steel scrap compris-

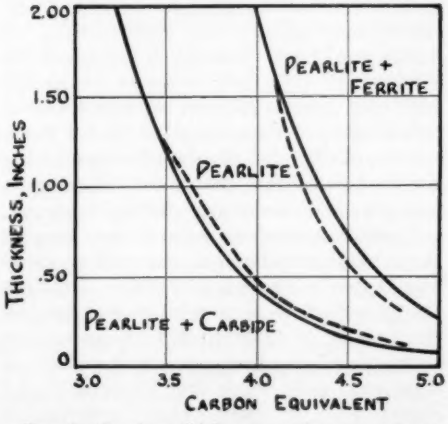

FIG. 2.—Section thickness, carbon equivalent, and microstructure of cast iron, according to Schneidewind and McElwee (*5*).

ing the charge and no pig iron included, followed by superheating in an electric furnace with the added touch of desulphurization. The average strength rose from an initial 28,000 psi to 52,500 psi. Earlier, in 1929, McPherran (*2*) reported using 94 percent steel scrap, adding ferrosilicon and ferromanganese and obtaining an average tensile strength of 58,300 psi with close to 90 percent of his tests falling between 50,000 and 70,000. It should be noted, however, that 1 percent nickel was added to this iron in the ladle. Consideration of large scrap additions, especially if an alloy addition is included, necessitates reference to the paper of Coyle and Houston (*12*). With a charge of 75 percent steel scrap in the cupola and 25 percent iron scrap or pig iron, tensile strengths of 50,000 to 70,000 psi were obtained through the addition of 1 to 4.5 percent nickel. This, of course, is not inoculation (shortly to be discussed) but graphitizing pressure by nickel rather than silicon and the strengthening effect of nickel on the matrix.

The study of high melting temperatures per se had an erratic beginning. While Hailstone (*13*) in 1913 had found improved strength due to superheating to 1,425°C., others including Longmuir (*14*), the first to report such experiments (in 1903), disagreed in recording maximum strength when melting and pouring at an intermediate temperature below 1,400°C. Elliot (*15*) carried his studies farther, namely to 1,510°C., but stated emphatically the undesirability of higher superheat. Actually it remained for Piwowarsky (*16*), in a series of researches beginning about 1924 and extending over a period of more than 15 years, to set forth systematically the effects of superheating and to demonstrate that these effects upon strength, microstructure, chilling tendency, etc., are associated with the undercooling of the eutectic solidification. His work at times was carried to as high as 2,000°C. although the later tests were largely concentrated in the vicinity of 1,600–1,650°C. Constructive additions to this literature on the effect of superheat were made by Krynitsky and Saeger (*17*) in 1939. Their experiments covered the melting temperature range from 1,400 to 1,700°C., with pouring effected at 100 to 250°C. above the liquidus. Strength increased with increase in temperature with the best values reported for the iron of lowest carbon equivalent (3.91).

The significance of these strength improvements goes somewhat beyond the quoted values, taken principally from standard arbitration bars of about 1.2-inch section. The benefits persist in larger sections, even in plain irons up to more than 3 inches in thickness. For example, Piwowarsky and Szubinski (*18*) pointed out that raising the melting temperature from 1,250°C. to 1,500°C. produced a tensile strength increase of 50 percent and increasing the section from 0.8 inch to 3.2 inches reduced this increase by only half. Von Frankenberg (*19*) in a series of tests ranging from 1.8 to about 3.6 percent carbon studied the effect of temperature change over the range 1,300 to 1,600°C. in sections from about 1.6 to about 3.2 inches. He recorded practically constant numerical increases due to this rise in temperature, which meant higher percentage improvements for the lower initial strengths of the higher carbon alloys and the heavier sections.

It is necessary now to return to Fig. 2. Briefly, superheating moves both of the curves of Fig. 2 to the right thus increasing the carbon equivalent at which, for any given section thickness, free ferrite may be avoided and similarly increasing the carbon equivalent below which free cementite will be encountered. The latter feature is of great importance for the present purpose in that it means a greater tendency to chill and, owing to brittleness of chilled metal, inability to develop inherent higher strength. Of major importance, however, is the fining of the graphite due to supercooling, regarded by various investigators as the result of (a) complete solution of the graphite of the raw materials, (b) a higher gas content perhaps principally oxygen or dissolved iron oxide, and/or an iron silicate slime as proposed by v.Keil and coworkers (*20*) permeating the molten iron but eliminated in the course of superheating. Accompanying the fining of the graphite is a more or less pronounced eutectiform arrangement of the short flakes, the extent of which varies according to composition, section and thermal history; this

condition effects a further restraint upon the development of the full inherent tensile strength of any iron.

INOCULATION

The history of inoculation is not quite clear cut, involving some complexities such as the cross-currents of developments in several countries, delays in publication, etc. Mere mention of the possible effects of deoxidizers, or their use in inadequate or otherwise incorrect amounts or by ineffective methods, is no assurance of initiating this practice. Based upon published records and actual commercial practice, credit would seem to belong to Meehan (*21*) who treated iron sometime before February of 1922 with calcium silicide, producing thereby high-strength gray iron from molten metal that would otherwise have cast white. Sometime prior to the issuance of his patent in 1924 I remember being astonished at a test from a keel block received from Meehan recording 60,000 psi in the first examination at the U. S. Naval Gun Factory and a most competent analyst named Faust establishing the presence by wet chemical analysis of calcium in the order of 0.01 percent! On the other hand, there is also clearly fixed in my memory a visit in the early twenties to the foundry of the Wm. Sellers & Sons Co. in Philadelphia and there watching A. E. Outerbridge (apparently mentioned only by Moldenke (*22*)) add in the ladle 75 percent ferrosilicon to iron made with partial scrap charges, and listening to him proclaim the excellence of the product. I recall no tests, but if any appreciable amount of iron oxide or silicate was in that iron he may very well have been inoculating. This statement should naturally be viewed in the light of the fact that in cupola practice the oxygen content of the iron is not likely to be high unless the temperature is high or there is moisture in the charge materials.

Piwowarsky (*23*) tried to set forth in a long tabulation the sequence of developments in inoculation but failed to demonstrate that such were the results secured by each investigator listed. Some of the references appear to be merely a listing of possible deoxidizers without recognition of the limits of their activity or effectiveness. At about the time of his earlier references to what is *assuredly* inoculation, namely in 1929, he wrote (*24*), rather inadequately:

> Observing the favorable influence of the deoxidation of cast iron prior to its pouring, the author was able to show on the basis of a substantial, but as yet unpublished, experimental series that cast irons with between 68,000 and 74,000 psi[2] bend strength (39,000 to 42,000 psi tensile strength) were improved by the addition of between 0.05 and 0.15% silicon as ferrosilicon or alsimin, resulting in an increase in bend strength by about 6,000 to 8,000 psi, i.e. to about 74 to 82,000 psi. "Hot-produced" cast iron types require for the production of dense blow-hole-free castings always a final treatment with a little silicon, aluminum, titanium or vanadium.

Since that date our knowledge of this subject has been considerably expanded and refined. It is known for example that the first additions of vanadium and also chromium (*25*) effect a measure of deoxidation and modification of eutectiform graphite, subsequent amounts serving as alloying additions to the matrix. Many years of successful practice have shown that the most efficient inoculants are *rich* ferrosilicon (75 percent silicon or higher) and *complex agents* containing one or more of Ca, Ti, Zr, Li, Al, usually along with high silicon content and in some cases limited amounts of manganese. A further improvement, by Chandler (*26*), specifies a complex alloy of this type containing chromium in addition to manganese, thus simultaneously effecting mild alloying, strengthening of the pearlite and reducing the hardness gradient in heavy sections. Mexican graphite has also, since about 1937, found a small field of use as an inoculant but is mild in its effect.

Referring once again to Fig. 2, it has been amply proved by experiment and extensive practice that the use of adequate additions of effective inoculants moves the left curve farther to the left (somewhat farther than superheating moved it to the right) while affecting not at all the right hand curve of Fig. 2. Thus the usable zone of 100 percent pearlite is widened. It then becomes possible to realize the strengths that cannot be de-

[2] These values are the author's conversions.

veloped when chilling occurs; in other words, the full pearlite zone has been moved out into the cementite area making possible pearlitic structures with over 45,000 to 50,000 psi tensile strength. Not only is chill reduced but there is benefit from reduction or complete elimination of any eutectiform arrangement of the graphite without much coarsening of the refined graphite flakes. Somewhat differently expressed, the possibilities for the occurrence of chilled iron are reduced or confined to thinner sections; a section that exhibited eutectiform graphite is changed to one containing small random flakes and only one that after superheating was chilled will, after inoculation, display eutectiform graphite in a pearlitic matrix.

Discussion of inoculation would not be complete without reference to the work of Norbury and Morgan (*27*). These investigators stated that adding titanium to iron and subsequently oxidizing it, preferably by bubbling CO_2 through the melt, resulted in under-cooling and the formation of fine graphite. The cause was believed to lie in the *molten surface* of the *insoluble titanate* particles formed in this reaction. Subsequent treatment with hydrogen avoided the undercooling and produced coarse graphite, *as did also* the replacement of titanium by additions of silicon, calcium-silicide or aluminum which latter were stated to be inoculants that produce *solid* nuclei and therefore coarse graphite. It is somewhat difficult in view of operation at 1,350°C. (thus no superheat) to coordinate these observations with the results of present high-temperature melting followed by inoculation through deoxidizing additions. This is especially true in view of operation over many years and in many foundries with Meehan's practice of large additions of steel scrap, therefore high temperatures and the final addition of calcium silicide, with resultant high strength and random arrangement of graphite flakes that are *not* large, as well as the Drant-Kessler process producing similar results, also with high scrap charges but with final additions of complex finishing alloys to the low-carbon irons thereby produced. It is difficult not to conclude that there is a distinct difference between the Norbury-Morgan product and that resulting from superheating and inoculating if only because of the lack of eutectiform graphite in the former work undertaken without superheat and the lack of coarse graphite resulting from the latter method (i.e., current practice in the United States) following inoculation. Moreover, the functioning of titanium is not clear. Many pig irons now reaching the American market contain substantial quantities and this is even more true in Western Europe. Its behavior may differ according to its presence in pig and other raw materials and whether, if added to a molten bath, it is as a simple or complex alloy. Furthermore, combination with both oxygen and nitrogen needs to be considered. The presence of silicon in most titanium alloys thus far used for inoculation also needs reexamination.

In reviewing the voluminous literature and the unpublished records of foundry practice dealing with high-temperature melting in both cupola and electric furnace, followed by the effective use of inoculants, one cannot escape the impression of the need for experiment or theory that will fully cover the recorded data. Granted that many of the observations (to say nothing of the theories advanced) may be inadequate or faulty and that metallurgical investigators may be expected never to be in full accord, there is, despite the excellent work of v. Keil and coworkers (*20*) and of Boyles (*28*), obvious need for more complete coverage of the mechanism of solidification. Under what conditions as to raw materials, temperatures of melting and pouring, composition (especially carbon and silicon), details of inoculation and cooling rate does graphite separate on the one hand directly from the melt and on the other hand, form by carbide decomposition? What are the details of nucleation and growth of the graphite? Gillett in 1934 (*29*) expressed the opinion that much remains to be explained. Since then, Schafmeister (*30*) has produced what may prove to be adequate support for the hypothesis of v.Keil using stereoscopic photomicrography and detecting silicate films or coatings on coarse graphite particles separated by dissolving away the surrounding iron. Confirmation of these and other related researches is needed.

ALLOYING

Interwoven with the changes in production practice here described are the contributions to high strength of the alloying elements. They have been in use by the iron foundryman for 40 years or more, but their value became more thoroughly and precisely recognized when used in conjunction with improved melting, deoxidation and molding practices. The principal metals in current use for their contributions as alloying additions are chromium and vanadium with their influence on carbide formation and increased chill as well as temperature-stability and graphite-refinement, manganese (above about 0.3 percent) for its stabilization of pearlite, molybdenum for its mild carbide effect, strong graphite refinement, and contribution to temperature-stability, copper for its pearlite stabilization, mild restraint of chill and precipitation hardening and nickel for its strong chill reduction, graphite-refinement and strengthening of the matrix by large additions. As in steels, the alloying elements are almost always employed in conjunction with one another to secure in any desired measure the contribution obtainable from each at minimum total cost.

For the purpose of this discussion, interest centers in the strength values produced, even though the alloy irons are designed only in part for their strength and often in larger measure for their wear resistance, for their ability to sustain repeated thermal shock and for other special properties. A few selected examples are shown in Table 1. In some instances the tensile strength values are accompanied by computations of this property, made in accordance with the formula devised by Schneidewind and McElwee (*5*) in their last paper under joint authorship. Their successful formula for predetermination of tensile strength from chemical composition was adequately proven on unalloyed irons initially, then given greater significance by determining the increases due to alloy additions. Within the limits of avoiding chill, the percentage increases in strength for one percent of each element are:

Vanadium, 50 percent
Molybdenum, 45 percent
Chromium, 22.5 percent

TABLE 1.—STRENGTH OF ALLOY CAST IRONS
(Mn at 0.5–0.6 unless otherwise noted)

Group and No.	Chemical composition, percent							Tensile Strength, p.s.i.			Pour Temp. °F
	C	Si	Ni	Cr	Cu	Mo	V	Actual Avg.	Alloyed	If unalloyed	
A1	3.05	2.00	1.00	.40	—	.45	.10	62,000	64,600	40,600	2,860
2	3.10	1.95	.80	.20	—	.35	.10	54,400	55,400	39,600	
3	3.20	2.20	—	.20	.50	.25	.20	49,300	48,400	36,200	
4†	3.00	1.00	1.75	—	—	.60	.10	74,300	74,600	48,300	Inoculated
5	3.30	1.60	.80	.40	—	.55	.10	58,400			Inoculated
6	3.30	1.85	—	.30	.85	.30	.10	57,500			Inoculated
B1*	3.20	2.10	—	.55	—	.55	—	56,800			
2	2.85	2.40	.65	—	—	.70	—	68,800	61,000	42,000	2,700
3	3.45	1.70	.80	—	—	.85	—	50,200			
4	2.70	1.90	1.50	—	—	1.30	—	70,400			2,710
5†	2.50	1.90	1.00	—	—	1.05	—	82,800	85,500	52,300	
6	3.00	2.00	—	—	1.00	.55	—	70,800			Elec. fce. iron
C1†	2.70	2.35	.85	.15	—	1.00	—	70,000			
2	2.90	2.10	.85	—	—	.50	—	60,000			
3*	2.95	2.10	1.75	—	.15	.80	—	65,000			
4*	3.00	2.00	—	—	—	.70	.15	55,000			
5	2.90	1.35	1.75	.35	—	—	—	88,200‡			
6	2.60	1.75	1.50	—	—	.60	.12	65,800			
7	3.00	1.30	1.00	.50	—	.70	—	59,400			

* With 0.70–0.80 Mn.
† With 0.90–1.05 Mn.
‡ Quenched and tempered.
A—Vanadium Corporation Records; B—Climax Molybdenum Publications; C—A.F.A. Alloy Cast Irons.

Nickel, 6.5 percent
Copper, 8.0 percent
Manganese, 4.0 percent

For irons that as cast possess an acicular structure in the matrix, such as B5 in Table 1, computed and actual values are in agreement only after low-temperature tempering to convert residual austenite to bainite. These acicular irons with their fine graphite have contents of carbon, silicon, nickel, molybdenum, and sometimes chromium that are adjusted to section size, required chill and shrinkage limitations. They are not as readily machined as pearlitic irons but as cast in green sand up to about 2 inch sections show tensile strengths of 70,000 to 85,000 psi and after transforming retained austenite by tempering at about 325°C., 80,000 to 95,000 psi. These acicular structures are a reminder that alloying elements judiciously adjusted to carbon-silicon-cooling rate criteria are being employed to alter the rate of transformation of austenite and that this alteration may be carried somewhat beyond that required for strengthening while retaining the pearlitic and readily machinable matrix.

Heat treatment of alloyed iron is still practiced to a considerable degree, yet overall apparently to a lesser extent than formerly. When it is used, it is largely limited to a stress-relief by subcritical tempering or to full annealing. Except for malleable and nodular irons, quenching and tempering and practiced to only a minor degree even though as shown by the single example of Table 1, very high strengths may be developed. Among the irons still being employed in the quenched and tempered (or stress-relieved) condition in light sections, mention needs be made of an acicular alloy composition of 3.25 percent carbon and 2.00 percent silicon with 1.10 percent chromium, 0.50 percent molybdenum and 0.30 percent nickel. The structure is naturally heavy with carbides and of importance for wear-resistant parts. Values as high as 100,000 psi are not too difficult to secure in the various quenched and tempered nonductile irons, but the products then lack toughness. The decline in this practice in the fields thus far discussed may well be related to the desire for some measure of ductility in cast irons of very high strength in view of the more or less stressed conditions resulting from the methods necessary to the attainment of this high strength and hardness.

A quick recast of the development of strength in gray irons of predominantly pearlitic matrix structure may be had by glancing at Fig. 3 which indicates solely the direction of movement of the cementite-pearlite boundary and of the zone of mottled structure resulting from superheating, change of cooling rate, inoculation, and alloying. It affords merely a quick qualitative indication, taking no account of these influences upon the size, shape, quantity, and distribution of the graphite particles which determine the degree to which the matrix structure is restrained from full development of its inherent strength.

MALLEABLE IRONS

Thus far consideration has been given only to improvements in the strength of cast iron, using the value in tension as the indicator. For many engineering uses, however, ductility of metals in addition to strength is of great importance, not only where simple tensile loading is concerned but even more critically with multiaxial stresses and sudden applications of stress. The ability to deform under occasional overload, rather than to rupture suddenly without deformation, is a frequent requirement. Many uses are satisfied, however, by limited elongation in the tension test and many metallic parts with such qualifications have rendered eminent service over long periods.

Brief cognizance must therefore be taken of several products falling within that portion of the cast iron carbon range that is in part or totally below 2.4 percent. First, of course, is malleable cast iron, cast white, with temper carbon formed through any one of many variants of annealing, having a chemical composition ordinarily within the wide limits of 2.25 to 3.10 carbon with silicon 0.70 to 1.20 and properties in tension ranging from a strength value of 53,000 psi minimum to 90,000 psi minimum with corresponding minimum elongations of from 18 to 2 percent. Minimum values of 60,000

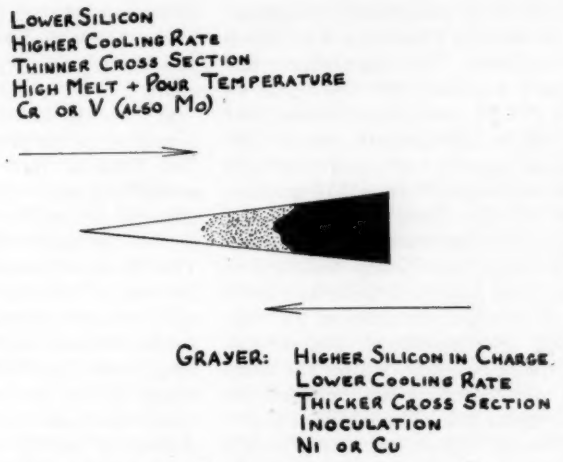

FIG. 3.—Basic variables in iron founding.

and 18 may be secured by suitable alloying, such as with copper plus molybdenum. The temper carbon aiding this development of ductility yet interfering with the full realization of the inherent strength of the steel matrix as in all cast irons, comprises more or less rounded graphite particles of irregular outline carrying innumerable projections that serve as points of stress concentration and initial fracture. Nevertheless, with their appreciable ductility, products within the range have enjoyed extensive use for very many years and have been the subject of much research, a large proportion of it devoted to shortening and cheapening the long cycle of heat treatment once universally required to effect the change from its white state as cast. In part this research has been very successful, but more in the direction of the higher strength varieties with their lower ductilities and pearlitic matrix. Owing to ease of fabrication, the softer grades have until now been produced in greater tonnage. Pressure from other ductile high-strength products is, however, enlarging the sphere of application of pearlitic malleable iron and may well be expected to attract the attention of users to an increasing degree. Driving minimum tensile strength to 100,000 psi will add interest even though the minimum elongation is quite small. As always, for any application, decision must be on the basis of engineering, metallurgy, and economics.

Over the past 25 years there have appeared now and then a goodly number of suggestions for irons that have been close relatives of malleable iron, not requiring more than short-time heat treatments or falling within the broad limits of chemical specifications for malleable irons, yet partaking of their properties. Some have enjoyed considerable industrial use and are worthy of brief mention. In a patent to McCarroll and Vennerholm (*31*) one finds an early version. With carbon just under the malleable iron range (1.90 to 2.30 percent) and silicon appreciably above (1.50 to 2.20 percent) the metal as cast is white but heat treatments of just a few hours duration precipitate about half of the carbon as temper carbon. With cooling rate after heating above the transformation range adjusted to suit exact composition and section size and reheating to yield a desired hardness, tensile strengths above 90,000 psi were obtained with presumably some ductility in view of the temper carbon. Further development of the idea by these investigators led to a lower-carbon content with silicon at about the same level or somewhat lower and a considerable copper content to insure good

castability and easy machining; chromium was added to develop a measure of hardness and wear resistance. The composition has been variously reported (*32*) (*33*) (*34*) as carbon 1.30 to 1.70, with silicon spread over the range 0.85 to 2.50 percent, copper 1.00 to 4.00 percent (usually 1.75), and chromium 0.30 to 2.00 but usually 0.40 to 0.75 percent. Of interest is the claim that with the higher silicon, precipitation hardening by copper may be avoided. Large tonnages of this product were made and there was much discussion of a proper designation, i.e. copper-malleable or copper-steel, but temper carbon is present in quantity after the usual short-time step-annealing and entitles the metal to inclusion here. The tensile strength has been variously reported between 105,000 and 120,000 psi with elongation ordinarily 1 or 2 percent but frequently higher.

From the same very active research group has come an additional malleable iron variety but in this instance within malleable iron limits for both carbon and silicon. It is produced by trace additions of bismuth and boron and is reported (*35*, *36*) to extend the useful range of malleable iron that may be cast completely white, from about the standard 2 inches to at least 4 inches. Here is an interesting modification of an early discovery now finding practical employment.

IRON WITH NODULAR GRAPHITE

The discovery of methods for producing iron castings with all or nearly all of the graphite or free carbon in spherolitic or nodular form has introduced into the field a contender of major stature. Whether discovery was accidental or the result of deliberate and painstaking search is unimportant; the contributions of Morrogh (*37*) and of Millis, Gagnebin, and Pilling (*38*) have opened a new vista in the field of ferrous metallurgy. Not that the transition from the old to the new was abrupt, for there are clear evidences in the work of v.Keil and associates (*20*), Krynitsky and Saeger (*39*) and Adey (*40*) that this end point was being approached and there were those such as Bolton (*41*), Gillett (*29*), Meyersberg (*42*) and possibly many others who had expressed a fervent hope that the graphite of cast iron would one day be precipitated or caused to grow in perfect spheroidal form. Nevertheless the accomplishment came as a pleasant jolt and represents notable achievement.

It is not the intention to review here many details of the properties of nodular cast iron. The interest that these properties have stimulated has resulted in a flood of experimentation, production, and publication beyond the possibilities of quick assimilation. The literature, however, is all so recent as to be readily located. Suffice it to point out that strength values, as cast, in plain and mildly alloyed iron with spheroidal graphite range from about 60,000 to 120,000 psi with elongations in the tension test from (in inverse order) about 3 to well over 20 percent. Annealing has the effect of greatly narrowing this range as will shortly be set forth while quenching and tempering indicates the possibility of attaining values well above 150,000 psi with appreciable elongation.

Methods of production, as is well known, have revolved about the patented practices of adding cerium to irons low in sulphur or of magnesium in sufficient amount to remove sulphur above about 0.02 percent while leaving an adequate amount after loss by volatilization of some magnesium to provide a residual of about 0.04 percent minimum. Effective use has been made of the two metals in association. For best results, i.e. to yield graphite completely or nearly completely in spheroid form along with a machinable matrix, silicon in the form of a suitable alloy such as 75 percent ferro-silicon is added just before pouring into molds. The amount of silicon added has been the subject of much discussion, especially in view of somewhat lower machinability at higher silicon contents in the iron, but the details of mechanical properties up to about 4 percent silicon have been clearly set forth by Schneidewind and Wilder (*43*) who have pointed to the increasing strength and decreasing elongation as silicon rises but that within the selected range, existing specification limits can be met without difficulty.

However, as concerns iron composition and alloying, much opinion that is not sound and many data established on inadequate coverage or too limited foundations have been published. No more striking example

exists than in the case of copper. Recent publications have covered territory from sweeping conclusions based on amounts of addition far beyond those in use, to precise establishment of the method of matrix structure control by this element (*44*, *45*, *46*, *47*, *48*, *49*). It is time to realize that copper may either have no influence on the properties of nodular iron or it may make definite contribution, according to the amounts added, the properties desired, any heat treatment that may be used and what is most important, the level of manganese in the iron. The elements copper and manganese both stabilize pearlite so that with manganese at 0.50 percent copper should be low. By low is meant about 0.30 percent, which is far above what will be obtained by ample use of a magnesium addition agent with 5 percent copper, normal copper content in scrap and foundry returns at the very maximum. If manganese is lower, copper may be much higher with a moderately favorable effect upon as cast strength and slight, if any, decrease in ductility. On the other hand, manganese and copper may both be raised, with or without other alloying elements added, to produce nodular iron in the tensile strength range of 85,000 to 115,000 psi with the expectable shift in elongation. And such irons may be annealed to produce the usual lower strengths (about 65,000 psi) and higher elongations if desired. The details are shortly to be published. The sole requirement is understanding use of two mutually dependent elements.

Titanium in nodular iron may have been equally maligned and improperly used. The metal has been reported (*44*) as very damaging to the properties of ferritic nodular iron. Work in the laboratories of the author's company confirms this provided the amount of titanium is quite high as, for example, in the range of 0.05 to 0.10 percent. The metal is being introduced through many varieties of pig iron but not enough studies have been undertaken to detect differences, if any, between titanium added as pig, as iron-rich alloys or silicon-rich alloys. Likewise insufficient study has been devoted to smaller percentages of titanium which are more common in nodular iron. Nor has adequate attention been given to the simultaneous presence of varying amounts of nitrogen. It is quite true that cerium corrects the ill effects of the amounts of titanium referred to above as detrimental, but it is necessary to know whether smaller amounts of this corrective may be adequate, or even none at all, required.

At this point it seems proper to refer to what may be termed seminodular irons—those that have part of their graphite as nodules and the remainder as stubby flakes or other partly converted forms. Millis and co-workers (*38*) indicate such products from very low magnesium additions. A paper by Estes and Schneidewind (*50*) soon to be published shows that irons in this category, which they call "upgraded irons," may be produced at various carbon equivalents (4.3 to 5.1 having been thoroughly examined) with tensile strengths of 50,000 to 60,000 psi and a nominal elongation of 5 percent, by the ladle injection of calcium carbide, magnesium oxide, and the rare earth oxides. Iron fully nodular resulted from substitution of magnesium metal powder for the oxides, as would be expected. Other intermediate cast irons, not quite so far advanced have already been mentioned and it is to be expected that still different approaches will be profitable. The pressure of possible decrease in cost should be considerable.

FUTURE POSSIBILITIES

I have pointed to past accomplishments in the development of high strength in one of the lowest cost metal products—a readily castable material of varied uses. What are the prospects for and directions of further change or improvement? I have indicated a few—there are surely many more. To recall those mentioned and comment upon them while introducing others should not be amiss.

First consideration should be given to the mechanism of solidification. Some of the important investigations have been noted; recent additions to these studies have come from the laboratories of Hultgren (*51*) and Morrogh (*52*, *53*). In a measure the methods are refinements of conventional procedures. One cannot help looking forward to the likelihood of some skillful and resourceful experimenter devising means for selective

behavior of a radioactive isotope to watch the progress of the solidification reactions. The foundryman may be led thereby to better control of small graphite flakes, of stubby flakes, of mixtures with imperfectly formed spheroids and perhaps the control of the number and size of perfect spheroids.

More investigative work should and will be done to evaluate the performance of titanium in irons, not only of itself but in conjunction with its origin and also its association with nitrogen.

Some work has been accomplished on the effects of aluminum in irons (*54*), but the investigations recorded, with a single exception, all deal with what may be termed alloying quantities. The oft-quoted limits of 0.03 percent maximum to avoid porosity when using green sand molds and 0.06 percent for dry sand molds needs reexamination. These values require questioning; under controlled production methods or association with some other metals, the limits may not hold and desirable results may accrue.

Improved methods of inoculation need study. The production of cheap calcium seems not far away, and it is necessary to know whether present alloys are most effective or whether introduction of commercially pure or very pure metal by methods effecting reaction only in the molten iron and with its oxygen and nitrogen, will present improvement. Similar studies with other alkali and alkaline earth metals may be profitable.

Additionally, attention should be directed to sulphur, for the general sulphur problem as it affects the entire ferrous metallurgical industry is of great importance to the economy of nodular iron processes as now known and practiced. Lowering sulphur makes possible savings through lower magnesium or cerium additions or both. Contributions in this direction have been made by means of reactions with alkali compounds, lime-rich slags in the electric furnace, ladle reactions with calcium-aluminate slags, ladle injection of calcium carbide and application of new cupola designs providing for basic linings, hot blast and controlled slags. Slag control holds out hope for reduction in the quantity of present metallic additions and may eliminate them; it should be worthwhile to read again the paper by v.Keil and associates (*20*) already mentioned.

Finally, I return to the comment on perfect spheroids. Quenching and tempering of nodular iron are now an accepted practice. The limitation in reaching for very high strength is low ductility or none at all, even brittleness. If the smaller the nodules the more perfect their form, it may be possible mathematically to estimate the most desirable size distribution and spacing. If control can be effective, I'd guess that close to 200,000 psi tensile strength would be possible and that with measurable elongation. This should be the ultimate in the strength of cast iron.

The great Albert Sauveur in a discussion of one of Dr. Burgess's papers (*55*) commented that there was need for the definiteness and finality of the latter's conclusions. It is, therefore, fitting to close with two quotations, the first of which is from that paper:

> May we not state the axiom that in a basic industry if there is a factor, . . . which is generally felt to play a capital role, limiting in an as yet undetermined way, the quality, output or cost of product; then, in such cases, all reasonable effort should be devoted to ascertaining the effects of the factor in question.

The second is from one of those many illuminating reviews of Dr. Burgess's successor, H. W. Gillett (*29*):

> Whether a theory that leads us to try out something . . . is a correct theory or not is of little moment if the trial leads to a solution of the difficulty.
>
> Yet we must hope for and search for an entirely correct theory, for were such . . . evolved and proven, one could use it with certainty to lead . . . to an immediate solution . . . rather than merely as a guide for experimentation in each recurring case.

Much too much time and energy have been spent in hammering away at these "recurring cases" in the field of cast iron. A great amount of information has already been accumulated. It is time that the chaff be tossed aside and some strenuous effort devoted to a search for a few more kernels of grain.

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HYDROGRAPHY.—*A comparison of the environmental characteristics of some shelf areas of eastern United States*.

FRANCIS E. ELLIOTT, WILLIAM H. MYERS, and WILLIS L. TRESSLER, U. S. Navy Hydrographic Office.

(Received April 27, 1955)

Recently surveys have been conducted of certain portions of the continental shelf along the eastern coast of the United States. The present paper attempts to bring together certain over-all aspects of the environmental features of this region. The parameters considered include physiography, temperatures, salinities, currents, sea level, transparency, color, climatology, and fouling.

PHYSIOGRAPHY

The continental shelf is part of the continent rather than a part of the oceans. It is the authors' belief that for a better understanding of the shelf, one can not divorce it from the coast and its hinterland. It is therefore suggested that the continental shelf, and the adjacent coastal lands to the first major break in physiography, be treated as belonging to the same physiographic province. In the northern part of a belt extending from Boston to Charleston, S. C., the land includes the Seaboard Lowlands, and in the southern part the Coastal Plain as defined by Fenneman. This seems entirely justified if one remembers that during the last glaciation, most of the continental shelf was emerged, and it seems reasonable to assume that the gross features of the relief were formed subaerially and only the microrelief was formed under submarine conditions, because only comparatively little time has elapsed since the return of the sea. From a geomorphological point of view one can divide this belt into a northern glacial part and a southern alluvial part. The boundary between the two runs in a westerly direction through Staten Island.

Going into more detail, in the northern part one can make further subdivisions from its position with respect to the ice sheet. The configuration of the shelf near Boston is typical of an eroded ground moraine dominated by drumlins. One should also expect to find other glacial features such as eskers, roches moutonnées, etc., however, bathymetric charts are not good enough at present to recognize these features. The Narragansett Bay area (Fig. 1, *a*, *b*) is located at the margins of glaciation and its dominant features are a succession of terminal moraines divided by a narrow outwash plain. The Harbor Hill moraines seem to extend from Long Island, at Orient Point, through Fishers Island to the shore of Rhode Island, while the Ronkonkoma moraine extends from Montauk Point through Block Island probably to Cuttyhunk Island, etc. The latter one particularly

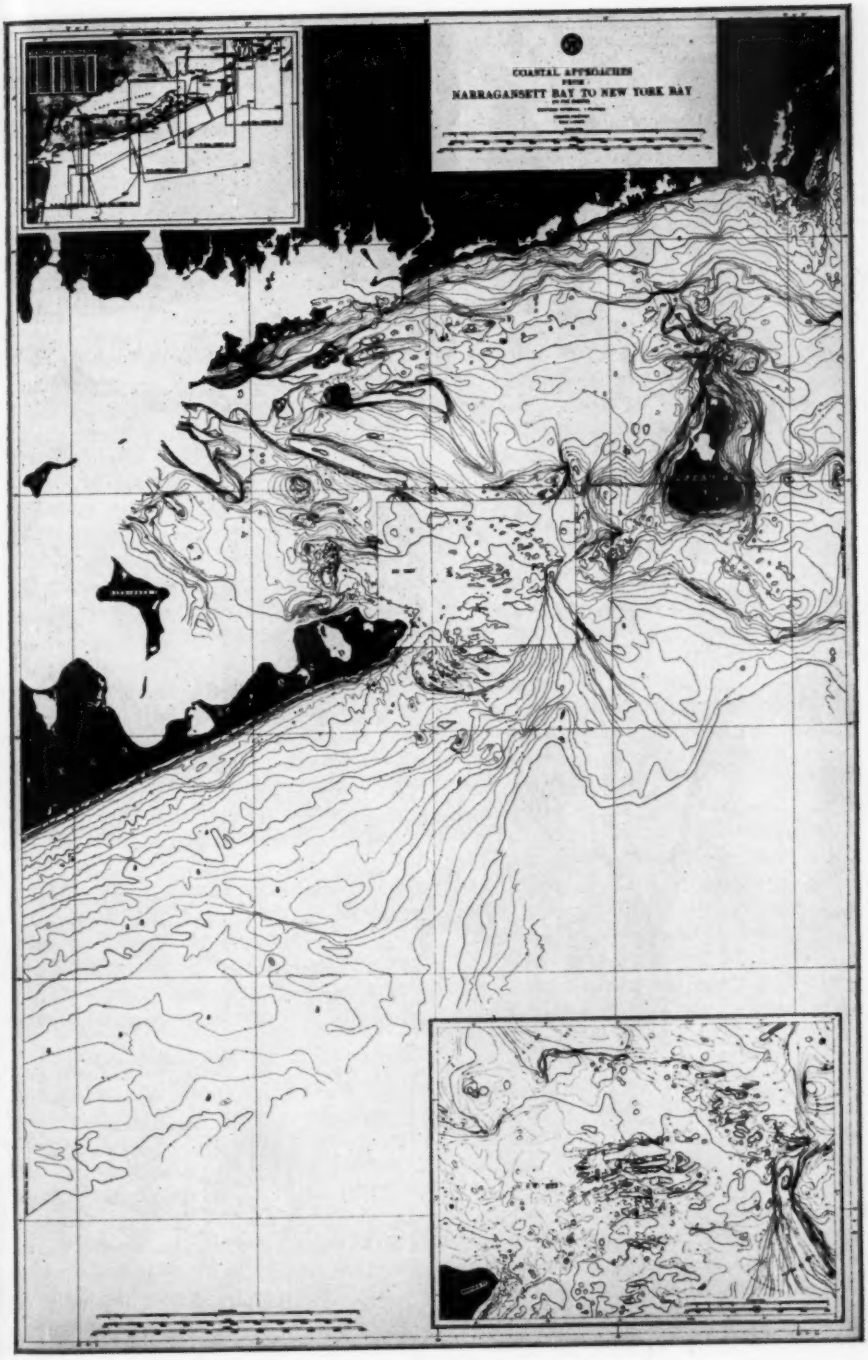


FIG. 1a

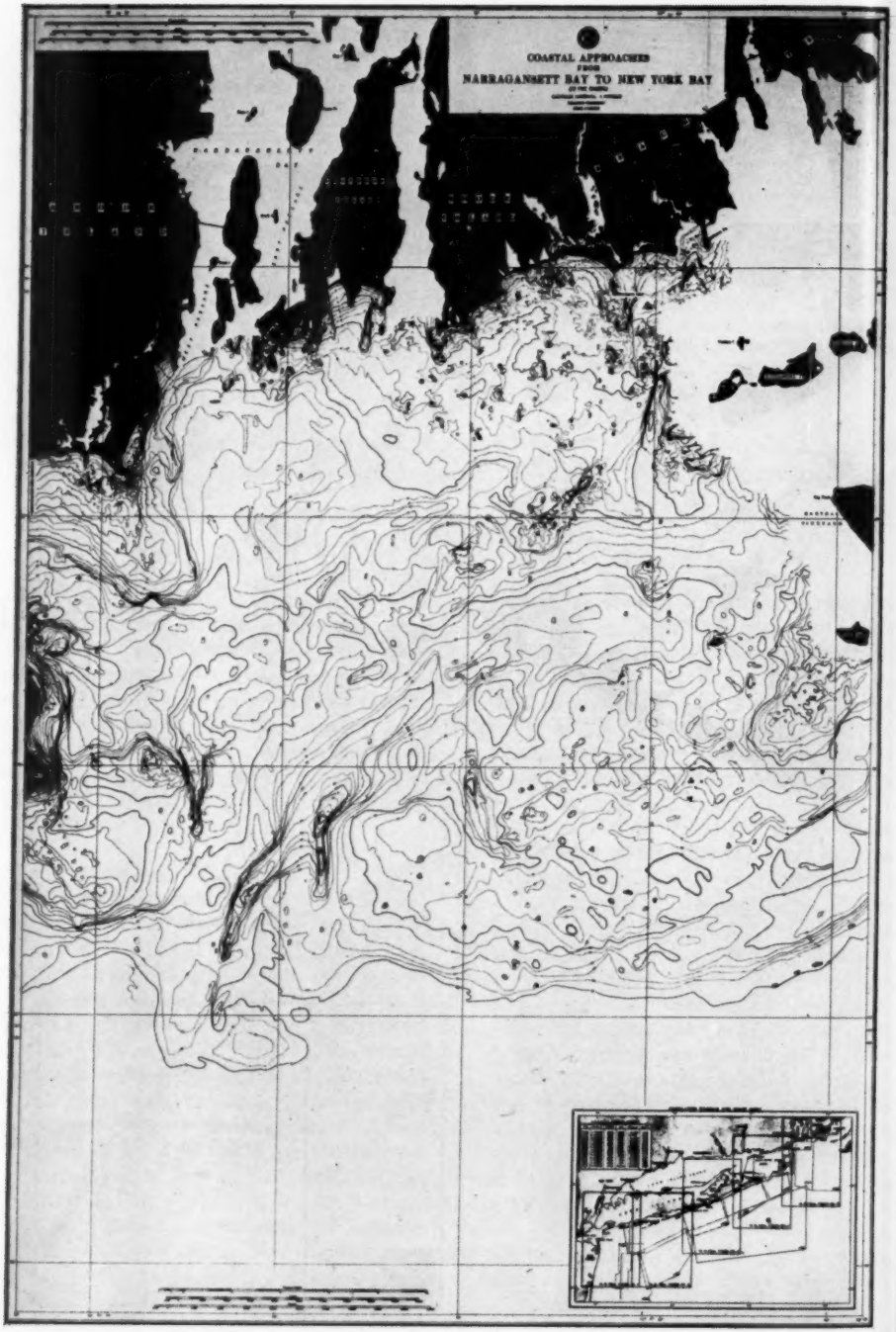


FIG. 1b

is easily recognizable on the bathymetric chart by its knob and kettle topography. The comparatively smooth areas between the moraines probably represent the outwash plain. Some of the deep holes south of Fishers Island may very well be kettles.

The approaches to New York (Fig. 2, *a*, *b*, *c*) are just beyond the ice limits and represent the outwash plain. The topography becomes considerably smoother and rolling. The elongated narrow ridges south of Jones Beach cannot be adequately explained; they may be submerged sand dunes or submerged off-shore bars which have encroached on each other. The Hudson has eroded a channel through the unconsolidated material, probably before the sea returned.

The southern part of our coastal belt is much less diversified than the northern part and a clear cut distinction between the individual areas can not be made.

The Delaware River, like the Hudson, has cut a channel which in this case is recognizable to about 15 fathoms where it ends at a ridge, which could be a submerged coastal terrace. The same terrace, if it is one, can be seen off Chesapeake Bay at about the same depth (Fig. 3). A Susquehanna channel, on the other hand, cannot be traced. The assumption that this ridge represents a former coastal terrace is not unreasonable in the light of the fact that there is a succession of such terraces on the adjacent embayed coastal plain.

Off Charleston no terrace appears within the limits of the chart (Fig. 4). However, numerous shallow depressions exist whose origin is unknown to the authors. It is suggested, however, that they may be a submerged continuation of the Carolina bays.

TEMPERATURE

Along the shelf from Narragansett Bay to Chesapeake Bay, in winter, the surface water isotherms run approximately parallel to the shoreline with temperature increasing in a seaward direction. At equal distances from land along this area of the shelf there is a latitudinal increase of temperature of only 1.5°–3°C. The water column from surface to bottom is isothermal in all localities. The isothermal condition may occasionally be disrupted by indrafts from river discharges. Such an interruption of the winter pattern might conceivably be felt off the Chesapeake Bay during January when there is an increased river discharge.

Spring warming begins any time between the middle of February and the middle of March and becomes more rapid after late April or early May. During May the latitudinal gradient along the shelf from Narragansett Bay to Chesapeake Bay increases to about 6°C. at the surface and about 4°–5°C. on the bottom. In the areas for which there is record, the temperature profiles show an inversion during late spring at varying levels in each locality. Off Martha's Vineyard and Montauk Point in the Narragansett Bay area, the inversion occurs between 20 and 40 meters, the only difference between the two locations being the slightly higher temperatures at Montauk Point. Off New York the inversion occurs between 30 and 40 meters and off Delaware Bay at approximately 40 meters. Immediately off the entrance to Chesapeake Bay the inversion is not at all prominent, but farther northward along the Virginia coast at Winterquarter Lightship, the inversion becomes evident at approximately 40 meters.

In June, surface temperature rises rapidly. The following values are reported across the shelf in the first to second week of the month: Off Martha's Vineyard and New York, 12–14°C.; off Delaware Bay, 16–18°C; and off Chesapeake Bay, 19–20°C. Maximum surface temperatures generally occur throughout the area early in August. Thermoclines which have been developing since late spring generally become steepest just before autumnal cooling. The upper 10–15 meters at this time are almost homogeneous with the thermocline generally present between the 15 and 30 meter levels. During the summer, the thermocline extends downward about 20 meters off Delaware Bay, which is the location of the thickest thermocline in the areas under consideration; in the other areas its spread includes usually not more than 10 meters. The thermocline off Martha's Vineyard is not as pronounced as in the other areas; here a temperature inversion occurs between 40 and 60 meters.

With autumnal cooling the thermocline

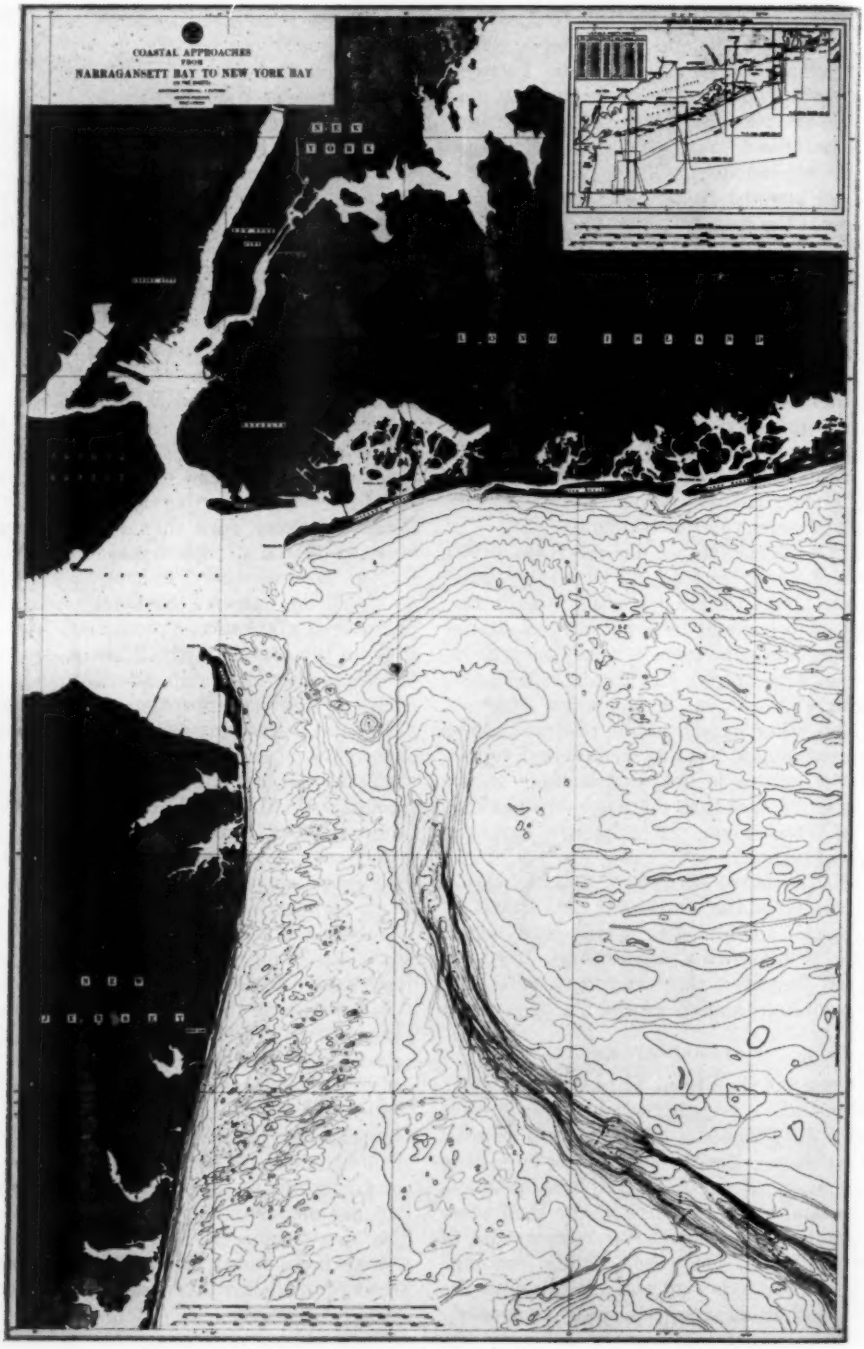


FIG. 2a

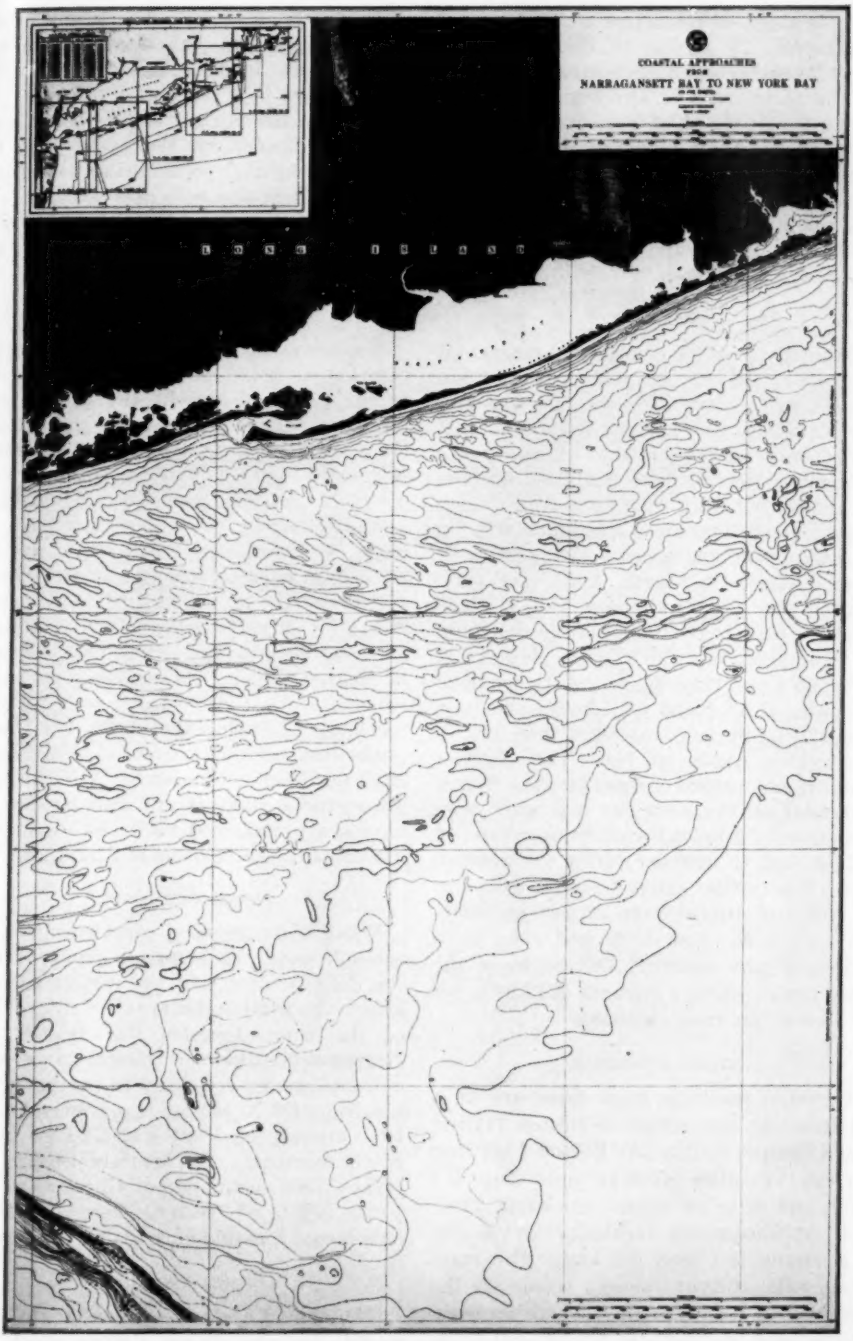


FIG. 2b

sinks to about 40–60 meters in all areas and is generally 20 meters in thickness. The upper layers become more homogeneous, and by the third week of October the surface waters across the shelf have cooled to 14°–15°C. off Martha's Vineyard, 15°–17°C. off New York, and 18°–19°C. off Chesapeake Bay. With decreasing temperature, by mid-December, mean temperatures are about 5°–8°C. along shore and 10°–13° C. along the outer edge of the shelf.

SALINITY

Salinity patterns along the continental shelf show greater seasonal stability than temperature patterns of corresponding periods. The basic salinity pattern is generally only altered close to shore by the freshening effect of river water and offshore by the salting from indrafts of more saline slope waters. The most striking change in the salinity pattern appears during the peak river discharge in March, April, and May. During this period minimum values which might be expected 8–10 miles from shore are: off Montauk Point, 30.8‰; off New York, 27‰; off Cape May, 30.5‰; off the coast of Virginia, 31.3‰; and off Chesapeake Bay, about 27‰. Surface water less than 32‰ is greatest in width off New York Harbor where it may extend offshore between 90 and 100 miles; off Delaware Bay it is never more than about 50 miles broad. Surface salinity values tend to increase during the autumn while the vertical range tends to decrease. In July and August steep vertical gradients may occur off New York and even more strikingly may occur off the mouth of the Chesapeake, where a gradient of 12.24‰ per 20 meters has been recorded.

TIDAL CURRENTS

Currents resulting from tides are at a minimum in the offings of Boston Harbor and Chesapeake Bay. At Boston Lightship current velocities average less than 0.1 knot, setting in an almost clockwise direction. At Chesapeake Lightship the velocity is generally less than 0.2 knot. The maximum tidal current velocity occurs off the entrance to Delaware Bay, where an average velocity of 1.4 knots occurring 0 hours after Greenwich transit, setting in a northwesterly direction, has been determined. Here, as at Boston, the current shifts continuously in a clockwise direction. At Charleston the current velocity is nearly the same throughout the cycle, with the direction shifting continuously clockwise. Off New York the velocities are nearly as weak as at Chesapeake Bay, but the current rotates clockwise as at Charleston and Boston.

NONTIDAL CURRENTS

A seasonal change in direction of nontidal currents is noted at Boston. During the winter (January–March) the predominant direction is ESE, in late spring and summer (May–July) the predominant direction is NE, and in fall (September–November) the direction is approximately N. Average nontidal current velocity is less than 0.1 knot. Off Narragansett Bay there seems to be no seasonal periodicity; the maximum velocity occurs in March with an average of 0.2 knot in a southwesterly direction. Off New York the major flow is to the south or southeast. Off Delaware Bay the major current set is to the south or southwest, seemingly following the shore. Off Chesapeake Bay, nontidal current directions vary from northeast to due south. At Charleston the nontidal currents are distinguished by having a pronounced northeast set in July with an average velocity of 0.22 knot.

WIND CURRENTS

Wind currents vary considerably in all areas; however, all areas are similar in that the deflective force of the earth's rotation causes the wind currents to set to the right of the wind direction. Off Boston and Narragansett Bay the directions of wind currents are not restricted to any particular quadrant. Off New York, wind current directions are restricted to the ESE to SW sector of the compass, off Delaware Bay to the NE to WSW sector, and off Chesapeake Bay to the NE to SW sector. There is no available record for wind currents off Charleston.

SEA LEVEL

The annual cycle of sea level is the same in all localities except Boston. Boston is

FIG. 2c

APPROACHES TO
CHESAPEAKE BAY

unique in that the maximum sea level height occurs in June, while maximum height does not occur until September in the other areas. The range of mean sea level increases southward. At Boston the range is .34 feet; in the Narragansett area, .35 feet; in the New York area, .60 feet; and in the Chesapeake area, .61 feet. The annual means of sea level along the coast show a constant rise from 1928. All areas except New York show approximately the same pattern for the years 1928–1950. The exception noted at New York occurred in the years 1938–1940 when sea level remained practically static, whereas in the other areas there was a general rise.

TRANSPARENCY

Transparency of shelf water appears to be correlated with salinity; i.e., areas of low salinity concentration are generally low in transparency. The low salinity water probably represents river effluent which is made turbid by suspended pollutants and sediments or large phytoplankton populations. In all areas transparencies are lowest within the bays, especially near the mouths of their tributaries. From the entrances of the bays shelf water becomes increasingly transparent in a seaward direction. This appears to be consistently true except in the offing of Narragansett Bay where a confused pattern of distribution results from the complex current running in and out of Long Island Sound, Narragansett Bay, Buzzards Bay, and Vineyard Sound.

WATER COLOR

The only complete coverage of water color in the area under consideration is in the region off Narragansett Bay and New York. The shelf water is predominantly blue-green, but local differences occur near land areas. Off Narragansett Bay and surrounding Montauk Point the water is yellow-green; the eastern and southern coasts of Block Island are surrounded by an area of light-green water, while the northwestern coast is surrounded by yellow-green water. Extensive areas of yellow-green water also appear off the coasts of western and central Long Island. A broad area of brown-green water extends across the entrance to New York Bay, and two similar areas of brown-green water are found off the western and south-western coasts of Block Island.

CLIMATOLOGY

Climatological characteristics of the six shelf areas are fairly similar and generally show differences only in degree. Mean annual temperature range decreases from Boston to Charleston except at New York. The precipitation pattern indicates that the greatest amount of rainfall occurs during the summer months, generally July and August, at most stations. Mean annual rainfall is greatest at Charleston and least at Delaware Bay and Block Island. Wind speeds are greatest at New York and off Narragansett Bay, where they may average as much as 16 knots. There is a seasonal variation in all areas with wind velocities lowest during the summer months and highest during the winter. Heavy fogs occur most frequently in the Narragansett Bay area and least frequently at Charleston. During the colder months of the year heavy fogs are most frequent off New York, Chesapeake Bay, and Charleston. In the other areas they are most frequent during the warm months. Appreciable swell occurs similarly in all areas, being least during July and August and greatest in late fall and winter.

FOULING

Fouling in Boston Harbor is very variable from year to year, the chief fouling organisms being hydroids, barnacles, bivalves, and tunicates. Other organisms occurring occasionally are tube worms and bryozoans (both filamentous and encrusting). Maximum growth of hydroids occurs in the Boston area in August. In Narragansett Bay, limited data indicate that fouling organisms are mainly bivalves, bryozoans, and algae with barnacles occurring at only one locality. Fouling has been estimated to occur on bottom objects in the entrance to Narragansett Bay at the rate of .05 pounds wet weight per square foot per month. Surface objects in the entrance to the bay accumulate fouling organisms at the rate of between 0.1 and 0.3 pounds per square foot per month. The fouling rate decreases in Narragansett Bay and in all other areas of

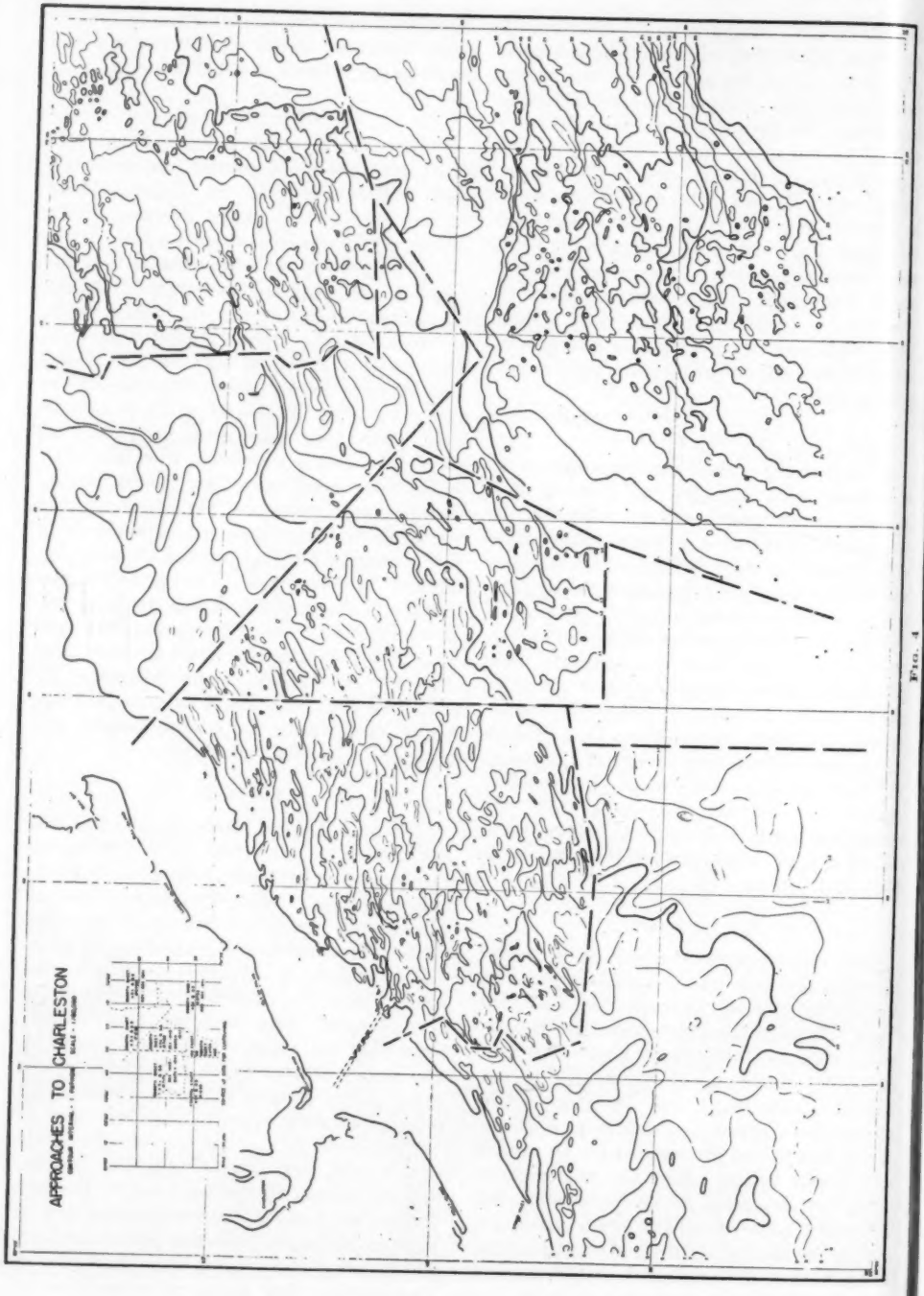


FIG. 4

the shelf as one proceeds seaward. Fouling in lower New York Harbor is light with barnacles and hydroids the principal organisms, and only traces of other forms. Conditions are similar from year to year but vary considerably within the area. Upper New York Harbor is only lightly fouled, with barnacles the prevalent form. Hydroid fouling is very light and no other forms are of any significance. Setting and growth of barnacles progresses during early spring and summer; once attached they may remain for several years. The growth of barnacles average about one-half inch in thickness of encrustation during the first four months of attachment. Rates of fouling of both surface and bottom submerged objects show a marked increase shoreward directly opposite the entrance to New York Harbor. In the Delaware, and Chesapeake areas, barnacle fouling appears to become less serious, and fouling by such organisms as mussels, bivalves and hydroids assumes greater importance. Barnacles again become of importance as fouling organisms at Charleston. On the north side of Delaware Bay, heavy fouling is caused by slime-producing diatoms, silt, and oysters, with barnacles present in small numbers only. Oysters disappear as fouling organisms on the ocean side of Cape May but the rest of the fouling picture is substantially that found in the bay proper. At Stone Harbor, north of Cape May on the ocean, mussels are heavy fouling agents and seem to be on the increase. In the Norfolk and Portsmouth areas of the Chesapeake Bay, fouling varies from moderate to heavy and is caused mainly by hydroids, barnacles and mussels. Bryozoans, sea-anemones, and tube worms also contribute to the fouling mass. Growth of the fouling mass takes place during the greater part of the year in this more southern region and at Charleston as well. Average fouling by mussels in the Norfolk area is approximately 1.5 pounds per square foot per year. Fouling at Charleston is caused mainly by hydroids and barnacles, while encrusting and filamentous bryozoa, marine worms, and miscellaneous nonboring mollusks occur more or less sporadically.

Marine borers appear to be a serious menace to wooden structures in all areas thus far examined except New York Harbor, where the damage caused is negligible. In Boston Harbor, borers are abundant and have reached an intensity of 100 per square inch in one month. In the Narragansett Bay region, borers are abundant at Block Island but are seen only occasionally at the Brenton Reef Lightship. Within the bay proper, there is a heavy infestation of marine borers. At Cape May at the entrance to Delaware Bay, borers are present to a limited extent only and then only during the period from May through July. The abundance of borers is very variable at different locations in the Chesapeake Bay area. At Norfolk, borers are abundant in some years and only moderate in others, while at Portsmouth, there seems to be a heavy infestation every year. Marine borers cause considerable damage at Charleston and are present in moderate to heavy amounts. The breeding season for borers in this region lasts from May through October or November.

The freedom which New York Harbor enjoys from marine borers is believed to be caused by the shape of the bay and the large inflow of freshwater from the Hudson River, both of which factors tend to decrease salinities and make an unfavorable environment for these organisms.

PALEONTOLOGY.—*Cenozoic pearls from the Atlantic Coastal Plain*. H. E. VOKES, The Johns Hopkins University, Baltimore, Md.

(Received April 28, 1955)

Recently (April 3, 1955) while on a field trip with an undergraduate class in geology from The Johns Hopkins University, I collected a large pair of valves of the pelecypod *Isognomon maxillata* (Deshayes) from the lower part of the Choptank formation in an exposure approximately one-half mile south of Kenwood Beach (Governor's Run), Calvert County, Md. Unfortunately, and characteristically, the specimen broke up almost as soon as collected, owing to the fact that, in life, the nacreous calcareous layers were apparently separated by unusually thick layers of organic conchiolin that now have decayed. As a result the shells tend to break into thin platy sheets as soon as removed from the enclosing matrix.

As I was about to discard the ruined specimen I noted an unusual structure lying in the matrix that had been adjacent to the left valve. When removed it proved to be a large, almost spherical pearl (Figs. 1, 2) whose outer layers, like those of the enclosing shell tended to exfoliate, apparently due to the destruction of the conchiolin. Fortunately the inner layers were more resistant and the specimen as finally secured has a greatest diameter of 10.2 mm; complete it is estimated that the diameter was somewhat in excess of 15 mm. The portion of the pearl that was adjacent to the shell is flattened by the breaking away or incomplete deposition of approximately five lamina of that portion of the pearl that is yet preserved. These lamina, as measured on their broken edges are 0.2 to 0.25 mm thick. It is probable that the pearl was originally attached to the inner side of the valve in this area after the manner quoted by Brown (1940, p. 369) from Kunz and Stephenson (1908, p. 57). The structure of the shell of *I. maxillata*, and the fact that most specimens from the Choptank formation show, even today, a considerable degree of mother-of-pearl luster, suggest that this specimen was originally a highly lustrous gem pearl.

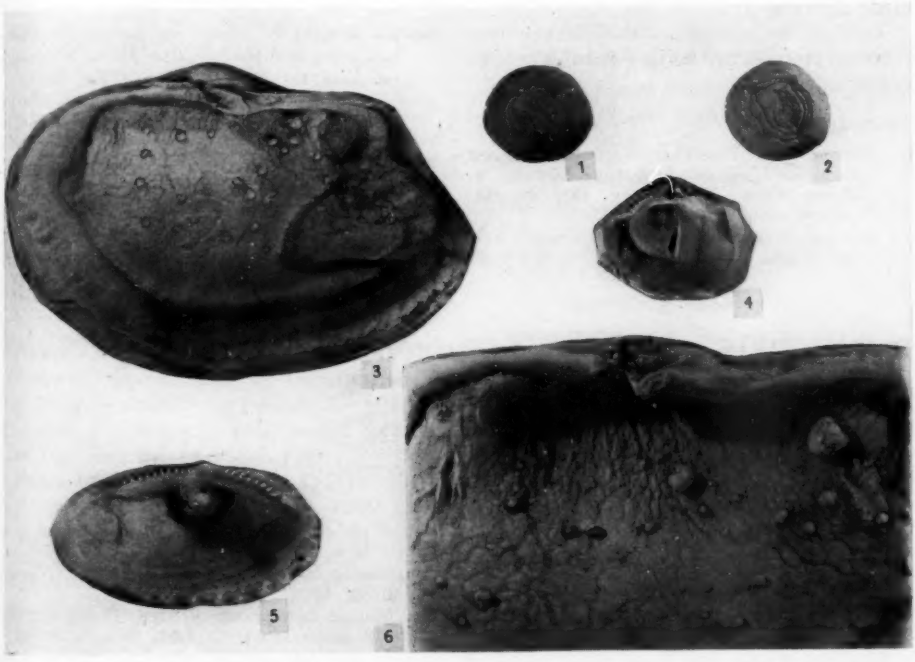
Brown, while describing a number of pearls from the Upper Cretaceous of Kansas, has tabulated (1940, table 1, p. 370) the reported occurrences of fossil pearls, which are known from all periods of the Mesozoic and Cenozoic eras. No definite pearls have been reported from the Paleozoic deposits, although Brown (p. 370) notes that minute pearl-like forms have been reported from the Upper Silurian. The absence of these structures from the Paleozoic seems almost certainly an accident of collecting, rather than an indication of the fact that such structures were not produced by Paleozoic pelecypods. More of the reported fossil pearls have been found in association with species of the family Isognomonidae which includes *Inoceramus* and *Isognomon* (= *Perna*, *Pedalion*, and *Melina*) than have been found in association with species referable to any other of the families of pelecypods. This family is represented in the upper Paleozoic by the genus *Bakevellia*. Six records listed by Brown are from species referable to the family Mytilidae (*Mytilus* and *Volsella*). This family is well represented in the Paleozoic and Newell (1942, p. 32) has shown that the related Myalinidae, abundant in the upper Paleozoic, had nacreous shells.

The specimen here described is the first "true" pearl to be reported from the American Miocene. Berry (1936, p. 464) has described a large "blister" pearl occurring in a specimen of *Panope americana* Conrad from the Choptank formation near Jones Wharf, Md., and Brown (1940, p. 367) has recorded a second occurrence in the same species from the same locality. A specimen in the collections of The Johns Hopkins University from the Choptank at Governor's Run, Md., has two smaller and very irregular pearls that are located immediately in front of the posterior adductor scar.

Species of the genus *Panope*, a burrowing pelecypod with a wide posterior gape where the large siphons emerge, seem to have been peculiarly subject to injury and to the possible entry between the shell and the

mantle of irritant material, probably sand grains. A pair of large valves of *P. floridana* Heilprin from the Caloosahatchie beds of the Florida Pliocene, in the Aldrich collection of The Johns Hopkins University show definite evidences of injury in the siphonal region of the shell, and on the inside of both valves there are a number of small, irregular pearls. There are 24 of these inside the right valve (Figs. 3, 6) and 22 in the left. The largest, in the right valve, has a diameter of 7.5 mm, one is 4.4 mm, and the rest are 3.5 mm or less. One of the smaller ones was ground in an effort to determine the structure; since *Panope* is a non-nacreous shell, the pearl, as to be expected, did not reveal well-developed laminar texture.

In addition to these specimens, the writer has collected two specimens from the upper Miocene Duplin marl at the Natural Well near Magnolia, N. C., that show "blister" pearls. One (Fig. 5), a left valve of *Glycymeris subovata* (Say), has a pearl of 6.5 mm greatest diameter located in the apex of the valve immediately below the umbo and behind the hinge-plate below which it projects slightly. There is no external evidence of injury or boring that penetrated the shell to account for the location of the "blister" in this part of the shell. The second specimen (Fig. 4) is a left valve of the common *Mulinia lateralis* (Say), the most abundant species in the fauna at the Natural Well. In the present specimen a large



FIGS. 1, 2.—Pearl from *Isognomon maxillata* (Deshayes): 1, Dorsal view; 2, Base showing some of the concentric laminae (× 1.5). The original specimen before exfoliation of the outer laminae was approximately the size of this illustration. Choptank formation, Governor's Run, Md.

FIGS. 3, 6.—*Panope floridana* Heilprin: 3, Right valve (× 0.5) with many irregular pearls (note the evident damage and repair to the posterior end of the valve); 6, oblique view of part of interior of shell (× 1); the large specimen in the upper right has a greatest diameter of 7.5 mm normal to the plane of the photograph. Caloosahatchie formation, Pliocene, Fla.

FIG. 4.—Blister pearl in interior of broken valve of *Mulinia lateralis* (Say) (× 1.2). Duplin marl, Miocene, Natural Well, N. C.

FIG. 5.—*Glycymeris subovata* (Say), oblique view showing pearl under the umbone and behind the hinge-plate (× 1.2). Duplin marl, Miocene, Natural Well, N. C.

"blister" 11.3 mm in greatest diameter, occupies much of the upper half of the interior of the valve between the adductor scars. The thickest development of the "blister is toward the posterior part of the structure, and coincides with a boring on the exterior of the valve, indicating that the initial irritant was probably an organism. However, a cut made in the "blister" revealed that it was hollow and contained a considerable amount of sand and mud, suggesting that the enlargement of the "blister" had been caused by this secondary irritant. Neither *Glycymeris* nor *Mulinia* has a nacreous shell, hence neither of these "pearls" like those of *Panope*, are true pearls in the commonly accepted sense of the term, although both types have a similar mode of origin.

To date the following "pearls" have been reported from the Atlantic coastal Cenozoic deposits:

"Blister Pearls":

Panope americana Conrad, two specimens from the Choptank formation, Miocene, at Jones Wharf, Md. (Berry, 1936, p. 464; Brown, 1940, p. 367).

Glycymeris subovata (Say), one specimen from the Duplin marl, Miocene, at Natural Well, N. C.

Mulinia lateralis (Say), one specimen from the Duplin marl, Miocene, at Natural Well, N. C.

"True or Spherical Pearls:

Isognomon maxillata (Deshayes), one specimen from the Choptank formation, Miocene, at Governor's Run, Md.

Panope americana Conrad, one specimen from the Choptank formation, Miocene, at Governor's Run, Md.

Panope floridana Heilprin, one pair of valves with numerous pearls, Caloosahatchie formation, Pliocene, Fla.

Anadara transversa (Say), one specimen from the Pleistocene at Wailes Bluff, Md. (Brown, 1946, p. 75).

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Kunz, George F., and Stevenson, Charles H. *The book of the pearl:* 548 pp. New York, 1908.

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ENTOMOLOGY.—*New names in the Homoptera.*[1] Z. P. Metcalf, North Carolina State College. (Communicated by H. Friedmann.)

(Received May 27, 1955)

The new names proposed herewith seem to me to be necessary for the reasons stated. The names are listed under the appropriate family, subfamily, and tribe, according to the classification which I now use in the card catalogue of the Homoptera of the World. This should enable any student to locate the forms concerned.

Primary homonyms have been replaced even where the species are no longer included in the same genera. All references have been checked against the originals.

[1]Contribution from the Entomology Department, North Carolina Agricultural Experiment Station, Raleigh, North Carolina. Published with the approval of the Director of Research as Paper No.642 of the Journal Series.

Family Cixiidae

Subfamily Cixiinae

Tribe Cixiini

Oliarus ovatus, n. n.

pro *Oliarus lacteipennis* Kusnezov, Ent. Nachr. **10:** 161. 1937.

nec *Oliarus lacteipennis* Fowler, Biologia Centrali-Americana **1:** 93. 1904.

Family Araeopidae (Delphacidae)

Subfamily Delphacinae

Tribe Delphacini

Nilaparvata caldwelli, n. n.

pro *Nilaparvata muiri* Caldwell and Martorell, Journ. Agr. Univ. Puerto Rico **34:** 193. 1951.

nec *Nilaparvata muiri* China, Ann. Mag. Nat. Hist. (9) **16:** 480. 1925.

Family DICTYOPHARIDAE

Subfamily DICTYOPHARINAE

Tribe DICTYOPHARINI

Dictyophara lindbergi, n. n.

pro *Fulgora acuminata* Lindberg, Comm. Biol. **10** (7): 106. 1948.
nec *Fulgora acuminata* Olivier, Encyclopédie méthodique . . . **6:** 571. 1791.

Family FULGORIDAE

Subfamily PHENACINAE

Levia, n. n.

pro *Helvia* Melichar, Wissenschaftliche Ergebnisse der Zweiten Deutschen Zentral-Afrika-Expedition 1910–1911 **1:** 123. 1912.
nec *Helvia* Stål, Bih. Svenska Vet.-Akad. Handl. **4:** 80. 1877.
Orthotype: *Helvia schubotzi* Melichar.

Family FLATIDAE

Subfamily FLATINAE

Tribe POEKILLOPTERINI

Poekilloptera walkeri, n. n.

pro *Poeciloptera producta* Walker, List of homopterous insects in the British Museum **2:** 452. 1851.
nec *Poeciloptera producta* Spinola, Ann. Soc. Ent. France **8:** 432. 1839.

Tribe NEPHESINI

Name to be restored: *Dalapax* Amyot and Serville.
nec *Pseudoflata* Guérin-Méneville.
Haplotype: *Flata postica* Spinola.

Dalapax was established by Amyot and Serville, Histoire naturelle des insectes. Hémiptères: 521. 1843, for the species *Flata postica* Spinola, Ann. Soc. Ent. France **8:** 420. 1839.

Pseudoflata was established by Guérin-Méneville, Iconographie du règne animal 1844: 360, as a subgenus of *Ricania* for *Pseudoflata nigricornis*, n. sp.

Nigricornis Guérin-Méneville is a synonym of *postica* Spinola.

Melichar, Ann. Nat. Hofmus. Wien **16:** 251. 1901, accepts the year 1838 as the date of the publication of Guérin-Méneville and therefore gives precedence to *Pseudoflata* over *Dalapax*. According to Hagen, Bibliotheca entomologica **1:** 309. 1862 and Horn and Schenkling, Index litteraturae entomologicae (1) **2:** 470. 1928, the entomological part of Guérin-Méneville's paper was not published until 1844. This date has also been accepted by Sherborn, Index animalium **21:** 5195. 1929; Neave, Nomenclator zoologicus **3:** 982. 1940; and Schulze, Kükenthal, and Heider, Nomenclator animalium **4** (21): 2907. 1935. Previously Marschall, Nomenclator zoologicus: 378. 1873, gave the date as 1846. Scudder, U. S. Nat. Mus. Bull. **19:** 266. 1882, also gave the date as 1846.

Panormenis melichari, n. n.

pro *Ormenis suturalis* Melichar, Wien. Ent. Zeit. **24:** 289. 1905.
nec *Ormenis striolata* var. *suturalis* Melichar, Ann. Nat. Hofmus. Wien **17:** 95. 1902.

Family ISSIDAE

Subfamily HEMISPHAERIINAE

Gergithus formosanus, n. n.

pro *Gergithus reticulatus* Matsumura, Trans. Sapporo Nat. Hist. Soc. **6:** 101. 1916.
nec *Hemisphaerius reticulatus* Distant, The fauna of British India **3:** 361. 1906.
nunc *Gergithus reticulatus* Distant.

Family CERCOPIDAE

Subfamily CERCOPINAE

Tribe CERCOPINI

Triecphorella kirschbaumi, n. n.

pro *Cercopis fasciata* Kirschbaum, Jahrb. Ver. Nat. Nassau **21–22:** 63. 1868.
nec *Cercopis fasciata* Fabricius, Mantissa insectorum **2:** 275. 1787.

Tribe EOSCARTINI

Eoscarta (Eoscarta) lombokensis, n. n.

pro *Eoscarta tristis* Jacobi, Zool. Jahrb. (Syst. Ökol.) **74:** 282. 1941.
nec *Eoscarta borealis* var. *tristis* Lallemand, Trans. Ent. Soc. London 1927: 112.

Keducarta walkeri, n. n.

pro *Triecphora antica* Walker, Journ. Linn. Soc. Zool. **10:** 289. 1870.
nec *Triecphora antica* Walker, List of homopterous insects in the British Museum **3:** 674. 1851.

Tribe COSMOSCARTINI

Cosmoscarta sundana, n. n.

pro *Cercopis liturata* Walker, Journ. Linn. Soc. Zool. **10:** 287. 1870.
nec *Cercopis liturata* Le Peletier and Serville, Olivier's Encyclopédie méthodique. . . **10:** 606. 1825.

Opistarsostethus walkeri, n. n.

pro *Cercopis dorsalis* Walker, Journ. Linn. Soc. Zool. **10:** 283. 1870.
nec *Cercopis dorsalis* Walker, List of homopterous insects in the British Museum **3:** 658. 1851.

Family APHROPHORIDAE

Subfamily APHROPHORINAE

Tribe PHILAENINI

Neophilaenus exclamationis var. **lindbergi**, n. n.

pro *Philaenus exclamationis* var. *nigerrimus* Lindberg, Not. Ent. **3**: 40. 1923.
nec *Philaenus exclamationis* var. *nigerrimus* Strobl, Mitt. Naturw. Ver. Steiermark **36**: 208. 1900.

Philaenus leucophthalmus var. **zetterstedti**, n. n.

pro *Cercopis spumaria* var. *obscura* Zetterstedt, Fauna insectorum Lapponica **1**: 515. 1828.
nec *Cercopis obscura* Fabricius, Entomologia systematica **4**: 49. 1794.

Tribe APHROPHORINI

Jophora compactilis, n. n.

pro *Aphrophora compacta* Matsumura, Annot. Zool. Japonenses **5**: 35. 1904.
nec *Aphrophora compacta* Walker, List of homopterous insects in the British Museum **3**: 701. 1851.

Family TETTIGELLIDAE

Subfamily TETTIGELLINAE

Tribe TETTIGELLINI

Amblyscarta crocea, n. n.

pro *Tettigonia aestuans* Walker, List of homopterous insects in the British Museum **3**: 750. 1851.
nec *Tettigonia aestuans* Fabricius, Entomologia systematica **4**: 20. 1794.

Amblyscarta frontaliana, n. n.

pro *Tettigonia frontalis* Germar, Mag. Ent. **4**: 64. 1821.
nec *Tettigonia frontalis* Donovan, Natural history of insects of China: [2]. 1798.

Amblyscarta nigrifascia var. **albidissima**, n. n.

pro *Tettigonia albida* Walker, List of homopterous insects in the British Museum **3**: 777. 1851.
nec *Tettigonia albida* Walker, List of homopterous insects in the British Museum **3**: 767. 1851.

Amblyscarta transversalis, n. n.

pro *Tettigonia transversa* Signoret, Ann. Soc. Ent. France (3) **1**: 342. 1853.
nec *Tettigonia transversa* Costa, Cenni Zoologici 1834: 89.

Dasmeusa flavescens, n. n.

pro *Tettigonia lurida* Signoret, Ann. Soc. Ent. France (3) **1**: 662. 1853.
nec *Tettigonia lurida* Germar, Mag. Ent. **4**: 70. 1821.

Poecilocarda nigripennis, n. n.

pro *Tettigonia scutellata* Signoret, Ann. Soc. Ent. France (3) **8**: 203. 1860.
nec *Tettigonia scutellata* Fabricius, Systema rhyngotorum **1803**: 44. 1803.

Tettigella alba, n. n.

pro *Tettigonia concinna* Walker, List of homopterous insects in the British Museum **3**: 755. 1851.
nec *Tettigonia concinna* Perty, Delectus animalium articulatorum **3**: 180. 1833.

Tettigella borneonensis, n. n.

pro *Tettigonia elongata* Walker, Journ. Proc. Linn. Soc. **1**: 167. 1857.
nec *Tettigonia elongata* Signoret, Ann. Soc. Ent. France (3) **2**: 495. 1854.

Tettigella chinensis n. n.

pro *Tettigonia trilineata* Melichar, Ann. Mus. Zool. St. Petersburg **7**: 132 (57). 1902.
nec *Tettigonia ruficeps* var. *trilineata* Fowler, Biologia Centrali-Americana **2**: 236. 1899.

Tettigella distanti, n. n.

pro *Tettigoniella cornelia* Distant, Ann. Mag. Nat. Hist. (8) **1**: 521. 1908.
nec *Tettigoniella cornelia* Distant, The fauna of British India **4**: 209. 1908.

Tettigella druryi, n. n.

pro *Tettigonia sanguinea* Westwood, Illustrations of exotic entomology **2**: 81, 1837.
nec *Tettigonia sanguinea* Fabricius, Systema rhyngotorum emendanda: 39. 1803.

Tettigella gabonensis, n. n.

pro *Tettigonia nigrolineata* Taschenberg, Zeitschr. Naturw. **57**: 446. 1884.
nec *Tettigonia nigrolineata* Herrich-Schäffer Deutschlands Insecten **164**: 17. 1838.

Tettigella montrouzieri, n. n.

pro *Tettigonia flavescens* Montrouzier, Ann. Soc. Agr. Lyon (2) **7**: 113. 1855.
nec *Tettigonia flavescens* Fabricius, Entomologia systematica **4**: 24, 1794.

Tettigella pallidicornis, n. n.

pro *Tettigonia festiva* Melichar, Homopteren-Fauna von Ceylon: 159. 1903.
nec *Tettigonia festiva* Fabricius, Systema rhyngotorum: 41. 1803.

Tribe GRAPHOCEPHALINI

Amahuaka angustula var. **fowleri**, n. n.

pro *Tettigonia angustula* var. *immaculata* Fowler, Biologia Centrali-Americana **2**: 292. 1900.

nec *Tettigonia immaculata* Walker, List of homopterous insects in the British Museum **3:** 740. 1851.

Astenogonia fabricii, n. n.

pro *Cicada bicolor* Fabricius, Systema rhyngotorum: 65. 1803.
nec *Cicada bicolor* Olivier, Encyclopédie méthodique . . . **5:** 748. 1790.

Conogonia ceramensis, n. n.

pro *Tettigonia tripunctata* Walker, Journ. Linn. Soc. Zool. **10:** 303. 1870.
nec *Tettigonia tripunctata* Fitch, Ann. Rep. State Cab. Nat. Hist. **4:** 55. 1851.

Epiacanthus guttiger var. **aurantius,** n. n.

pro *Tettigonia guttigera* var. *dispar* Horváth Term. Fuzétek **22:** 371. 1899.
nec *Tettigonia dispar* Germar, Mag. Ent. **4:** 71. 1821.

Graphocephala flavovittata, n. n.

pro *Tettigonia multicolor* Signoret, Ann. Soc. Ent. France (3) **1:** 363. 1853.
nec *Tettigonia multicolor* Walker, List of homopterous insects in the British Museum **3:** 760. 1851.

Family LEDRIDAE

Subfamily KOEBELINAE

Tribe THYMBRINI

Rhotidus kirkaldyi, n. n.

pro *Ledropsis stâli* Kirkaldy, Bull. Hawaiian Sugar Planters Assoc. Div. Ent. **3:** 26. 1907.
nec *Ledropsis stâli* Melichar, Homopteren-Fauna von Ceylon: 143. 1903.

Family EUSCELIDAE

Subfamily EUSCELINAE

Tribe EUSCELINI

Anoplotettix fuscovenosus var. **horváthi,** n. n.

pro *Thammotettix fuscovenosus* var. *inornatus* Horváth, Rev. Ent. **14:** 165. 1895.
nec *Thamnotettix inornatus* Van Duzee, Trans. Amer. Ent. Soc. **19:** 303. 1892.

Hesium falleni, n. n.

pro *Cicada biguttata* Fallen, Nya Handl. Svenska Vet.- Akad. **27:** 27. 1806.
nec *Cicada biguttata* Fabricius, Species insectorum **2:** 325. 1781.

Tribe ATHYSANINI

Athysanus argentarius, n. n.

pro *Cicada argentata* Fabricius, Entomologia systematica **4:** 38. 1794.
nec *Cicada argentata* Olivier, Encyclopédie méthodiqué . . . **5:** 759. 1790.

Athysanus fabricii, n. n.

pro *Cicada reticulata* Fabricius, Entomologia systematica **4:** 44. 1794.
nec *Cicada reticulata* Linnaeus, Systema naturae **1:** 436, 1758.

Nephotettix apicalis de Motschulsky

In Metcalf, Bull. Bernice P. Bishop Museum **189:** 126. 1946, I reported that I could not distinguish between *Nephotettix bipunctata* and the commonly accepted varieties *apicalis* de Motschulsky and *cincticeps* Uhler. Unfortunately, I did not discover that Fabricius (1803) described his species as *Cicada bipunctata*, Systema rhyngotorum: 78, which was preoccupied three times:

Cicada bipunctata Scopoli, Entomologie Carniolica: 115. 1763.
Cicada bipunctata Linnaeus, Systema naturae (ed. 12) **1** (2): 710. 1767.
Cicada bipunctata Gmelin, Caroli a Linné Systema naturae **1** (4): 2111. 1789.

The next available name for Fabricius's species is *Pediopsis apicalis* de Motschulsky, Étud. Ent. **8:** 110. 1859. The *bipunctata* of Fabricius and the *cincticeps* of Uhler, either as species or varieties, will become synonyms of *apicalis*.

Phlepsius bergi, n. n.

pro *Deltocephalus variegatus* Berg, An. Soc. Cient. Argentina **8:** 264. 1879.
nec *Deltocephalus variegatus* de Motschulsky, Étud. Ent. **8:** 112. 1859.

Remadosus osborni, n. n.

pro *Euscelis* (*Athysanus*) *magnus* var. *piceus* Osborn, Florida Ent. **6:** 20. 1922.
nec *Athysanus piceus* Scott, Ent. Monthly Mag. **12:** 97. 1875.

Tribe THAMNOTETTIXINI

Thamnotettix matsumuri, n. n.

pro *Thamnotettix acuminatus* Matsumura, Journ. Coll. Sci. Tokyo **23** (6): 27. 1908.
nec *Deltocephalus acuminatus* Uhler, Proc. Zool. Soc. London 1895: 80.
nunc *Thamnotettix acuminatus* Uhler.

Neobala boliviensis, n. n.

pro *Thamnotettix pallidus* Osborn, Ann. Carnegie Mus. **15:** 67. 1923.
nec *Thamnotettix karrooensis* var. *pallidus* Cogan, Ohio Journ. Sci. **16:** 192. 1916.

Family DELTOCEPHALIDAE

Subfamily DELTOCEPHALINAE

Tribe DELTOCEPHALINI

Deltocephalus amuriensis, n. n.

pro *Deltocephalus bilineatus* Lindberg, Comm. Biol. **3** (6): 7. 1929.
nec *Deltocephalus bilineatus* Gillette and Baker, Bull. Colorado Agr. Exp. Stat. **31:** 85. 1895.

Deltocephalus marginellanus, n. n.

pro *Deltocephalus marginellus* Osborn, Ann. Ent. Soc. Amer. **19:** 346. 1926.
nec *Deltocephalus marginellus* Osborn, Ann. Carnegie Mus. **15:** 41. 1923.

Deltocephalus obtusus, n. n.

pro *Deltocephalus simplex* Haupt, Bull. Palestine Agr. Exp. Stat. **8:** 29. 1927.
nec *Deltocephalus simplex* Van Duzee, Trans. Amer. Ent. Soc. **19:** 304. 1892.

Ederranus subangulatus, n. n.

pro *Cicada lutea* Sahlberg, Acta Soc. Sci. Fennicae **1:** 88. 1842.
nec *Cicada lutea* Olivier, Encyclopédie méthodique . . . **5:** 758. 1790.

Gillettiella labiata var. **gillettei,** n. n.

pro *Deltocephalus labiatus* var. *rufus* Gillette, Bull. Colorado Agr. Exp. Stat. **43:** 28. 1898.
nec *Deltocephalus abdominalis* var. *rufus* Sahlberg, Not. Fennica (n. s.) **9** (12): 329, 1871.

Jassargus (Jassargus) distinguendus var. **gallicus,** n. n.

pro *Deltocephalus distinguendus* var. *longiceps* Rey, Échange **10:** 46. 1894.
nec *Jassus* (*Deltocephalus*) *longiceps* Kirschbaum, Jahrb. Ver. Nat. Nassau **21–22:** 135. 1868.

Jassargus (Jassargus) distinguendus var. **reyi,** n. n.

pro *Deltocephalus distinguendus* var. *confinis* Rey, Échange **10:** 46. 1894.
nec *Deltocephalus confinis* Dahlbom, (1850) Handl. Svenska Vet.-Akad. 1850: 193.

Unoka gillettei, n. n.

pro *Athysanus ornatus* Gillette, Bull. Colorado Agr. Exp. Stat. **43:** 29. 1898.
nec *Athysanus ornatus* Perris, Ann. Soc. Linn. Lyon **4:** 174 (94). 1857.

Tribe XESTOCEPHALINI

Xestocephalus izzardi, n. n.

pro *Xestocephalus minutus* Izzard, Ann. Mag. Nat. Hist. (10) **17:** 598. 1936.
nec *Xestocephalus minutus* Distant, The fauna of British India **7:** 58. 1918.
= *Ootacamundus minutus* Distant.

Tribe CICADULINI

Cicadula capensis, n. n.

pro *Cicadula nigrifrons* Naudé, Ent. Me. Dept. Agr. Union of South Africa **4:** 86. 1926.
nec *Cicadula nigrifrons* Forbes, Report of the State entomologist of Illinois **14:** 67. 1885.

Tribe BALCLUTHINI

Balclutha haupti, n. n.

pro *Balclutha flava* Haupt, Bull. Palestine Agr. Exp. Stat. **8:** 37. 1927.
nec *Gnathodus impictus* var. *flavus* Baker, Can. Ent. **28:** 36, 38. 1896.

Family COELIDIIDAE

Subfamily TARTESSINAE

Tartessus evansi, n. n.

pro *Tartessus obscurus* Evans, Pap. Proc. Roy. Soc. Tasmania **1936:** 54. 1936.
nec *Tartessus obscurus* Schmidt, Stett. Ent. Zeit. **81:** 54. 1920.

Family AGALLIDAE

Subfamily AGALLINAE

Agallia lindbergi, n. n.

pro *Agallia insularis* Lindberg, Comm. Biol. **14** (1): 196. 1954.
nec *Agallia insularis* Berg, An. Soc. Cient. Argentina **17:** 39. 1884.

Family IASSIDAE

Subfamily IASSINAE

Tribe IASSINI

Batrachomorphus fabricii, n. n.

pro *Cicada prasina* Fabricius, Entomologia systematica **4:** 38, 1794.
nec *Cicada prasina* Pallas, Reise durch russischen Reichs **1:** 729. 1773.

Stragania matsumuri, n. n.

pro *Macropsis dorsalis* Matsumura, Journ. Coll. Agr. Sapporo **4** (7): 301. 1912.
nec *Macropsis dorsalis* Provancher, Petite faune entomologique du Canada **3:** 292. 1889.

Tribe KRISNINI

Krisna walkeri, n. n.

pro *Bythoscopus testaceus* Walker, Journ. Proc. Linn. Soc. **1:** 173. 1857.
nec *Bythoscopus testaceus* Walker, List of homopterous insects in the British Museum **4:** 1163. 1852.

Family IDIOCERIDAE

Idiocerus bakeri, n. n.

pro *Idiocerus trifasciatus* Osborn, Ann. Carnegie Mus. **15:** 19. 1923.

nec *Bythoscopus trifasciatus* Kirschbaum, Jahrb. Ver. Nat. Nassau **21–22:** 167. 1868.
nunc *Idiocerus trifasciatus* Kirschbaum.

Family TYPHLOCYBIDAE

Subfamily TYPHLOCYBINAE

Tribe EMPOASCINI

Empoasca canariensis, n. n.

pro *Empoasca unicolor* Lindberg, Comm. Biol. **6** (9): 8. 1936.
nec *Empoasca unicolor* Gillette, Proc. U. S. Nat. Mus. **20:** 731. 1898.

Empoasca martorelli, n. n.

pro *Empoasca incisa* Caldwell and Martorell, Journ. Agr. Univ. Puerto Rico **34:** 125. 1952.
nec *Empoasca incisa* Gillette, Proc. U. S. Nat. Mus. **20:** 735. 1898.

Tribe ERYTHRONEURINI

Erythroneura canariensis, n. n.

pro *Erythroneura affinis* Lindberg, Comm. Biol. **14** (1): 248. 1954.
nec *Erythroneura affinis* Fitch, Ann. Rept. State Cab. Nat. Hist. **4:** 63. 1851.

Family CICADIDAE

Subfamily TIBICENINAE

Tribe PLATYPLEURINI

Platypleura (Platypleura) schumacheri, n. n.

pro *Platypleura fenestrata* Schumacher, Zoologische und anthropologische Ergebnisse einer Forschungsreise **5:** 86. 1913.
nec *Platypleura fenestrata* Uhler, Proc. Acad. Nat. Sci. Philadelphia **13:** 282. 1861.

Tribe CYCLOCHILINI

Psaltoda plaga, Walker

Psaltoda plaga Walker, List of homopterous insects in the British Museum **1:** 109. 1850.
Will replace:
Cicada argentata Germar, Rev. Ent. Silbermann **2:** 66. 1834.
nec *Cicada argentata* Olivier, Encyclopédie Méthodique . . . **5:** 759, 1790.

Tribe TIBICENINI

Tibicen walkeri, n. n.

pro *Cicada marginalis* Walker, List of homopterous insects in the British Museum **4:** 1128. 1852.
nec *Cicada marginalis* Scopoli, Entomologie Carniïloca **1763:** 113. 1763.

Tribe FIDICININI

Fidicina africana, n. n.

pro *Cidcaa plebeja* Linnaeus, Systema naturae. (ed. 12) **1** (2): 707. 1767.
nec *Cicada plebeja* Scopoli, Entomologie Carniolica: 117. 1763.

Tribe DUNDUBINI

Terpnosia obscurana, n. n.

pro *Terpnosia obscura* Liu, Bull. Mus. Comp. Zool. **87:** 98. 1940.
nec *Terpnosia obscura* Kato, Bull. Cicadidae Mus. **2:** 3, 17. 1938.

Tribe PLATYLOMIINI

Platylomia kingvosana var. **viridescens,** n. n.

pro *Platylomia kingvosana* var. *virescens* Liu, Bull. Mus. Comp. Zool. **87:** 92. 1940.
nec *Platylomia virescens* Distant, Ann. Mag. Nat. Hist. (7) **15:** 66. 1905.

Subfamily CICADINAE

Tribe CICADINI

Cicada signoreti, n. n.

pro *Cicada punctipes* Signoret, Ann. Soc. Ent. France (3) **8:** 180. 1860.
nec *Cicada punctipes* Zetterstedt, Fauna insectorum Lapponica **1:** 525. 1828.

Family TIBICINIDAE

Subfamily TIBICININAE

Tribe TETTIGOMYINI

Xosopsaltria thunbergi, n. n.

pro *Tettigonia punctata* Thunberg, Dissertatio entomologica de hemipteris rostratis capensibus **1:** 7. 1822.
nec *Tettigonia punctata* Fabricius, Supplementum entomologiae systematicae **1798:** 516. 1798.

Family MEMBRACIDAE

Subfamily CENTROTINAE

Tribe LEPTOCENTRINI

Leptocentrus formosus, n. n.

pro *Leptocentrus formosanus* Kato, Trans. Nat. Hist. Soc. Formosa **18:** 32. 1928.
nec *Leptocentrus formosanus* Matsumura, Annot. Zool. Japonenses **8:** 15. 1912.

Tribe GARGARINI

Umfilianus gerstaeckeri, n. n.

pro *Centrotus fenestratus* Gerstaecker, Claus's Decken's Reisen in Ost-Afrika **3** (2): 429. 1873.
nec *Centrotus fenestratus* Thunberg, Dissertatio entomologica de hemipteris rostratis capensibus **1:** 3. 1822.

Tribe CENTROTINI

Centrotus cornutus var. **kirschbaumi,** n. n.

pro *Centrotus abbreviatus* Kirschbaum, Jahrb. Ver. Nat. Nassau **21–22:** 67. 1868.
nec *Centrotus abbreviatus* Fabricius, Systema rhyngotorum **1803:** 23. 1803.

LETTERS TO THE EDITOR

*A Consequence of Inequalities Proposed by Baker and Erickson**

The purpose of this note is to point out that the inequalities for Reiner-Rivlin fluids proposed by Baker and Ericksen[1] imply that the rate at which the stresses do work in distorting the fluid is always positive when the distortion matrix is nonzero. The latter condition is often imposed on theories of plasticity and fluid dynamics.

It is a trivial matter to verify that, for any quantities t_i and d_i $(i = 1, 2, 3)$,

$$3\sum_{i=1}^{3} t_i\Delta_i = (t_1 - t_2)(d_1 - d_2) + (t_2 - t_3)(d_2 - d_3) + (t_3 - t_1)(d_3 - d_1), \tag{1}$$

where $\Delta_i = d_i - \frac{1}{3}(d_1 + d_2 + d_3)$. In a Reiner-Rivlin fluid,[2] the stress matrix $\mathbf{T}$ is given in terms of the rate of deformation matrix $\mathbf{D}$ by

$$\mathbf{T} = f_0\mathbf{1} + f_1\mathbf{D} + f_2\mathbf{D}^2, \tag{2}$$

Here f_1 and f_2 are scalar invariants of $\mathbf{D}$, while f_0 is a scalar invariant of $\mathbf{D}$ if the fluid be compressible, an arbitrary hydrostatic pressure if it be incompressible. Hence the principal directions of $\mathbf{T}$ and $\mathbf{D}$ coincide and the principal values t_i of $\mathbf{T}$ are given in terms of the corresponding principal values d_i of $\mathbf{D}$ by

$$t_i = f_0 + f_1 d_i + f_2 d_i^2. \tag{3}$$

The inequalities considered require that $t_i > t_j$ whenever $d_i > d_j$, or, equivalently

$$(t_i - t_j)(d_i - d_j) > 0 \quad \text{whenever } d_j \neq d_j. \tag{4}$$

The distortion tensor $\boldsymbol{\Delta}$ is the deviator of $\mathbf{D}$, $\boldsymbol{\Delta} = \mathbf{D} - \frac{1}{3}\,\mathrm{tr}\,\mathbf{D}\,\mathbf{1}$, so its principal values are the quantities Δ_i. The rate at which the stresses do work in distorting the fluid is, by definition,

$$\mathrm{tr}\ \mathbf{T}\boldsymbol{\Delta} = \sum_{i=1}^{3} t_i\Delta_i. \tag{5}$$

If (4) holds, it follows immediately from (1) that (5) is positive unless $d_1 = d_2 = d_3 = 0$, in which case $\boldsymbol{\Delta} = 0$, which is the desired result.

One can construct counter-examples to show that (4) does not follow from the requirement that (5) be positive. For incompressible materials, we always have $\mathrm{tr}\,\mathbf{D} = 0$ so that $\boldsymbol{\Delta} = \mathbf{D}$. We then have, from (2),

$$\mathrm{tr}\ \mathbf{T}\boldsymbol{\Delta} = f_1\,tr\,\mathbf{D}^2 + f_2\,\mathrm{tr}\,\mathbf{D}^3.$$

Suppose that $f_1 = a$, $f_2 = b\ tr\ \mathbf{D}^2$, where a and b are positive constants. Then, since $\mathrm{tr}\,\mathbf{D}^2 \geqq 0$, the equality holding only if $\mathbf{D} = 0$, it is clear that (5) is positive. For incompressible materials, it is known[1] that (4) is equivalent to

$$f_1 - f_2\,d_i > 0 \text{ if } d_j \neq d_k \quad (i, j, k \neq),$$

$$f_1 - f_2\,d_i \geqq 0 \text{ if } d_j = d_k \quad (i, j, k \neq).$$

It is easily shown that the example considered here does not satisfy these conditions if, with a suitable choice of units, $d_1 = 2\sqrt{c}$, $d_2 = \sqrt{c + \frac{1}{4}} - \sqrt{c}$, $d_3 = -\sqrt{c + \frac{1}{4}} - \sqrt{c}$, where $c = a/b$.

J. L. ERICKSEN
Naval Research Laboratory

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[1] BAKER, M., and ERICKSEN J. L. *Inequalities restricting the form of the stress-deformation relations for isotropic elastic solids and Reiner-Rivlin fluids.* Journ. Washington Acad. Sci. **44**: 33–35. 1954.

[2] For a discussion of these fluids, see TRUESDELL, C. *The mechanical foundations of elasticity and fluid dynamics.* Journ. Rat. Mech. and Anal. **1**: 125–300. 1952.